METALLURGIA

The British Journal of Metals

(INCORPORATING THE METALLURGICAL ENGINEER)

Readers are invited to submit articles for publication in the editorial pages; photographs and/or drawings suitable for reproduction are especially welcome. Contributions are paid for at the usual rates. We accept no responsibility in connection with submitted manuscript. All editorial communications should be addressed to The Editor, "Metallurgia," 21, Albion Street, Manchester, 1.

PRINCIPAL CONTENTS IN THIS ISSUE:

	Page
The Future of the Light Metal Foundry Industry. By W. C. Devereux, F.R.Ae.S.	49–52
Estimation of Wetting Agent Concentration. By A. B. Ashton, M.Sc., and N. J. Stead *Surface active materials, or cutting agents, have, in recent years, found increasing application in electroplating and allied processes. Sometimes the results have been disappointing owing to the lack of closely controlled conditions. Making allowance for wastage has been a difficulty; for this purpose a reliable routine test is necessary and a rapid and simple test is described.*	53–56
Hardenability of Steel *Results are given of investigations on some aspects of isothermal transformation and end-quench hardenability; further developments of the end-quench hardenability test; a hardening test for low-carbon and shallow hardening tools; the effect of hardenability and the air hardenability of steels.*	56–58
Chromising Processes. By D. W. Rudorff *The process of diffusing chromium into iron and steel surfaces, the use of granular chromium or ferrochrome and of liquid chromium are briefly discussed. Particular attention is directed to the use of the gaseous phase which has made great progress in Russia and Germany.*	59–62
The Determination of the Hardness of Martensite and Austenite by means of the Micro-Hardness Tester. By H. Hanemann *The results are given of an investigation carried out to determine the hardness of martensite, austenite and their transition phases in a 1.7% carbon steel. A description is also given of a new micro-hardness tester used for the experiments.*	62–65
Aluminium Trains	66
Correspondence *The Shape of a Material's Reaction to Force.*	67–68
Microporosity in Magnesium Alloy Castings. By W. A. Baker, B.Sc. *Microporosity is the principal problem in casting magnesium alloys and an investigation is described which has been carried out to determine the primary causes and to develop remedies. The author's conclusions and recommendations are given.*	69–70
Metallographic Identification of Ferro-Magnetic Phases. By Miss E. M. A. Harvey, M.A., B.Sc. *The "magnetic etchings" method for the metallographic identification of ferro-magnetic phases is perhaps not so widely known as it deserves to be. This brief description of the method and apparatus used is given so that workers may have available details of the simple apparatus.*	71–72
War-time Consumption of Non-ferrous Metals	73–74
High Tensile Steel for Castings. By W. West, C. C. Hodgson and H. O. Waring *Investigations have been carried out by the authors on several aspects of steel-making and foundry practice in connection with the production of medium and high tensile steel castings. These are discussed and the results indicate a high degree of success from their practical application.*	75–79
The Yielding Phenomenon of Metals. Part VI. By Georges Welter *Influence of speed and loading conditions.*	80–84
The Microchemical Balance. By David W. Wilson	85–88
Metallurgical Digests	89–96A

Subscription Rates throughout the World - - 24/- per annum, Post free

Published monthly by THE KENNEDY PRESS, LIMITED.

REGISTERED OFFICE: 21, Albion Street, Manchester, 1.
Telephone: Central 0098. *Telegrams:* "Kenpred," Manchester.

LONDON OFFICE: Bedford Street, W.C. 2.
Telephone: Temple Bar 2629.

ACCUMULATED EXPERIENCE

If there is any difficulty in connection with the use of the "4-6" type of Molybdenum High Speed Steel (now the only specification of this type of steel being made), we place our accumulated experience at your disposal.

As pioneers of this type of steel, and holders of British Letters Patent covering the "4-6" specification (No. 534304 and now at the disposal of the Ministry of Supply to meet government requirements), we have amassed a great deal of manufacturing experience and data regarding fabrication, treatment and usage of this steel—all of this we offer freely to users of the "4-6" type of steel, whether of our manufacture or not.

Users of Darwins MOREX Molybdenum High Speed Steel will, of course, need to make no change whatsoever in their methods—they have been using what is now universally adopted as the most efficient of the Molybdenum High Speed Steels, and the treatment and performance will, naturally be exactly as before.

DARWINS

MOREX

"4-6" type of Molybdenum-Tungsten

HIGH SPEED STEEL

Patent No. 534304.

TOLEDO

DARWINS-TOLEDO

FINE STEELS MADE BY CRAFTSMEN

DARWINS — **DARWINS LIMITED**

TOLEDO — **ANDREWS TOLEDO LIMITED**

FITZWILLIAM WORKS & TOLEDO STEELWORKS · SHEFFIELD

Export Division : DARWINS-TOLEDO OVERSEAS LIMITED, SHEFFIELD.

METALLURGIA

THE BRITISH JOURNAL OF METALS.

INCORPORATING "THE METALLURGICAL ENGINEER."

JUNE, 1945 VOL. XXXII. No. 188

The Future of the Light Metal Foundry Industry

THE future of a section of industry cannot really be considered except against the broader background of our whole industrial future. The light alloy foundry industry has no separate future, nor has the foundry industry nor even the metal industry. All are part of the global total of British Industry and, moreover, are strongly influenced by by industry and trade throughout the world.

The demands of war have made heavy inroads into our national resources and for the first time in our history we have become a debtor nation, and that at a time when we have undertaken heavy obligations—social services, increased standards of living and full employment, social security, education reforms, etc.—and have incurred heavy debts to other countries. To meet these formidable obligations will call for a tremendous industrial effort to increase our productivity from our limited resources of manpower and capital. Increased productivity can be achieved only by increased efficiency—by an increase in the output of wealth per man-hour.

TO meet the exigiencies of war the aluminium industry has effected a seven-fold expansion. New factories were erected and equipped and existing ones greatly extended; in addition, a valuable personnel has been trained to cope with production for Service requirements. It is of vital importance that these factories and their personnel should be applied to peace-time developments in order to provide outlets for their products. With this in view the Aluminium Development Association staged an exhibition at Messrs. Selfridges, London, available to the public during the whole of this month, at which all sections of the aluminium industry are represented. The variety of exhibits clearly show the remarkable diversity of uses to which this metal or its alloys can be applied.

In view of the importance of the subject it is opportune that the Eighth Edward Williams Lecture before the Institute of British Foundrymen at the annual general meeting, held in London on June 16, 1945, should deal particularly with the light metal foundry industry. Many aspects are discussed and it has only been possible to present here some of the views given by the lecturer,

W. C. DEVEREUX, F.R.Ae.S., M.I.A.E.,

Chairman and Managing Director, High Duty Alloys Ltd.
Managing Director, Magnesium Castings and Products Ltd.

It has been shown in recent studies that the average American worker produces in an hour about twice as much wealth as the average British worker. If we are to compete in international trade—and that is the only way we can obtain the essential imports of food and raw materials we need to maintain ourselves—we must achieve a comparable productivity or lower our standard of living.

During the war the light alloy foundry industry, in common with all other industries, has shown how productivity can be increased by long-range planning and by the enthusiastic co-operation of management and labour in pursuit of a clear-cut objective. These two factors must be continued after the war. It is imperative to arouse in the people realisation of the importance of increased productivity, of the fact that on it will depend not only full employment and social security but also the maintenance of our position as a leading industrial country and even the way of life of the individual citizen. Production must be stimulated by long-range planning—by the formulation of broad objectives which will spread confidence and secure continuity of production.

After studying conditions in the light alloy foundry, industries both here and in America, Mr. Devereux considered our principal high production foundries at least as efficient as their opposite numbers in the States. He emphasised, however, that their efficiency is proportional to the extent to which they can rely on orders for sufficient numbers of casting to enable them to construct first class patterns, dies, and jigs and to utilise fully their machines and handling equipment.

These very efficient foundries, which have been organised and equipped by our sacrifices during the war should rank as important capital assets in our national economy and it is essential to ascertain how they will fit into our post-war industrial structure in order that the most efficient of them if not all of them, shall remain intact; not only the plant and machinery but also the fine teams of managers, technicians and craftsmen who, by their achievements, have shown their ability to organise and to improve the quantity and quality of output.

The requirements of a peace-time economy will probably be far more complex than those of a war economy. There will be a far greater variety of capital and consumer goods to be made; many of these will be required in quantities unsuitable for the use of mass production methods, therefore, a revival of the jobbing foundry to produce these goods in short runs can be expected. The actual balance between man and machine in this jobbing work will depend very largely on the numbers of young people who will enter the industry to learn its crafts and practices.

Light Alloy Foundry Achievements during the War

The output of aluminium castings in the period 1924 to 1933 remained fairly steady at roughly 10,000 tons a year. In 1934-35 a fifty per cent. increase in output brought this figure up to 15,600 tons according to the fifth Census of Production. Output rose at an increasing rate from that time on until the outbreak of war brought

about a rapid expansion of output, which, it is estimate, reached about 28,000 tons in 1940, 32,000 tons in 1941, 47,000 tons in 1942, 56,000 tons in 1943, reaching a peak rate of output of about 65,000 tons early in 1944.

A very rough estimate of the structure of the industry at this peak output stage would give about 25,000 tons as the output of three firms producing over 5,000 tons a year each, about 15 or 16 thousand tons as the output of seven or eight firms producing between 1,000 and 5,000 tons a year and roughly 20,000 tons produced mainly by about 130 light alloy foundries with the residue made up by some 500 firms producing occasional aluminium castings.

The magnesium casting industry in this country is of more recent origin, but, according to recent figures, the output of these castings had reached a figure of 400 tons a year by 1936. The war has brought about an expansion in the production of magnesium even more spectacular in proportion than in the production of aluminium castings. The production of die cast incendiary bombs at a peak rate of 4,327,000 castings per month account for a large proportion of the production of magnesium castings representing about 700 tons a month.

Post-War Use of Plant and Equipment

The bulk of the equipment now available is designed chiefly for the production of aircraft parts, that is mainly for small and medium sized castings. Such equipment is well suited to the quantity production of equipment, fittings and furnishings for he rehousing programme, for automobile construction, etc. The equipment capablea of handling larger castings will be an asset for the ship-building, railway, and mechanical engineering fields.

Among the larger firms, i.e., those with an output of 5,000 and more tons per annum no optimum pattern can be found. One large group consists of several foundries making castings of aluminium, magnesium, iron, foundry equipment and wrought light alloy products. Another very large group of foundries is part of a much larger organisation whose principal interest lies in the manufacture of aero engines and motor cars. Yet another organisation is devoted entirely to the manufacture of light alloy castings while others are concerned equally with the manufacture of wrought light alloys.

If this is complicated then the structure of the smaller foundries is even more so. They range from the jobbing foundry of a large general engineering concern through a variety of foundrics devoted entirely to the business of casting light alloys, to the highly specialised foundries producing nothing but die-cast aluminium pistons.

There are more than 600 foundries in the country making aluminium castings to some extent and yet about 70 per cent. of the total output of aluminium castirgs comes from about a dozen firms. Owing to the much more complex processes involved in casting magnesium alloys, this branch of the industry naturally tends to be more compact.

For these and other reasons Mr. Devereux confined most of his remarks to conditions as they apply to the larger units of the industry although, of course, much that concerns these large foundries will apply to the smaller units as well. In the limited time available he could not give a complete picture of the ways in which the light alloy industry has achieved its wartime expansion. As an example of the way in which mechanisation has been employed to obtain very high outputs, despite severe shortage of skilled workers, he cited the foundries of the Rolls Royce Company at Glasgow, and, by means of a series of illustrations, showed what can be done with first class equipment used in an intelligent manner. Many other fine foundry layouts have been built in recent years, but time did not permit particular reference to them, Mr. Devereux, however, emphasised that, in a foundry, modern conceptions of building, equipment, and layout can provide pleasant and clean working conditions.

Prices and Future Markets

The question of price takes a paramount place in discussion of the extent and nature of future markets for light alloy castings. As is generally known the price of primary aluminium has now been reduced from £110 to £85 per ton and still further reductions can be made if the import from low-cost producers is not interfered with.

Discussing the effects which this reduction and the improvements in production methods and equipment will produce in the price of engineering light alloy components, Mr. Devereux found himself in a somewhat difficult position. He is strongly opposed to the system of quoting for castings in terms of a price per pound. Few engineers would expect to buy fully machined engine parts on a weight basis nor would they expect their machines to sell by the pound. The difficulty, therefore, is to present prices in general terms without using the price per pound basis. In order to facilitate generalised comparisons, however, he used a weight basis emphasising that they should not be taken as signifying any concession to the system as applied to specific castings. In discussing this question of prices the fact that, during the past six years, all costs have risen well above pre-war levels will be appreciated. One of the best examples is the crankcase casting of the Rolls Royce Merlin engine. A major proportion of these crankcases have been produced in a foundry built and equipped for full scale mass production during the war. Production has been at the rate of 500 per week. When aluminium was at a cotnrolled price of £110 per ton, or 11·8 pence per pound, this casting was produced for 2s. 5d. per pound. It is a complicated casting made in a high-duty, heat-treated alloy—Hiduminium R.R.50 With the present ruling price of £85 per ton, or 9·1d. per pound for aluminium, the casting is being made for below 2s. 3d. per pound.

Those figures refer to castings made from primary materials. Great advances have been made in the recovery of high quality secondary alloys from scrap. Secondary aluminium of high quality can be made to very close composition tolerances for less than £45 per ton, or 4·8d. per pound. This means that such a casting as this could be made for less than 1s. 7d. per pound.

A step further is the possibility of making a rather less complicated casting by the gravity die casting process. An example of an important casting made by this method is the bottom half crankcase of the Merlin engine. This casting is produced to the exacting requirements of aero-engine practice, with all the searching tests that implies, at a selling price of less than a shilling a pound.

The implications of this statement call for very deep and careful consideration. It means in effect that a similar casting of equal volume in a heavy metal, such

as cast iron, would have to sell for below 4d. per pound to equal the price of the light alloy casting, even if we omit from our considerations the further savings—and these are quite considerable—which result from the greater ease of machining the light alloy casting, its lower handling and transport costs and the premium due for its great saving in weight as a machine component. It follows that for large outputs of light alloy castings where the standards of quality are less exacting than those in force for aircraft components, still lower prices will obtain. In these circumstances, the industry can confidently look forward to a market for light alloy castings far greater and very much more varied than anything we knew before the war.

It must, of course, be appreciated that the above remarks apply only in the case of castings made on a continuous production basis in quantities sufficient to permit the use of all the labour saving devices which have become so highly developed during the war. The volume of orders must permit the use of the very best patterns and die equipment and the fullest utilisation of facilities for mechanised production and material handling methods.

The picture drawn is that of the large mass production foundry working at the peak of efficiency and concentrating chiefly on these applications which will allow the continuous demand for a relatively small variety of castings. A great number of designs, however, call for castings in which the matter of strength is of very little importance. In practically all of these the most important object is a low price for the finished goods and a pleasing appearance having sales appeal. In most cases lightness is also an advantage.

In this field there are great possibilities for large and medium scale production of pressure and gravity die casting of light alloy ; especially for magnesium alloys, although Mr. Devereux is of the opinion that the field will still be dominated by aluminium, including the secondary aluminium alloys, the production of which, to exacting standards of quality, has now become so highly developed.

Finally, Mr. Devereux is of the opinion that there is a growing appreciation, resulting from modern developments in casting and quality control, of the economies offered by castings compared with fabricated assemblies. Engineers have already shown their confidence in castings for highly stressed components in aircraft and they are becoming increasingly appreciative of the advantages which the flexibility of the casting process offers in design and of the freedom from " stress raisers " which the jointless cast component offers.

Alloys and Alloy Development

In his lecture, Mr. Devereux directed particular attention to alloys and alloy development, but it is noteworthy that among the striking features of light alloy development in this war has been the way in which alloys, well established before the war, have been able to meet the increasing demands placed upon them, the pressing into service of alloys which were in the development stage at the outbreak of war, and the up-grading of medium strength alloys, particularly those made from secondary materials.

One alloy, developed in 1927 for a particularly high stressed aero-engine, has continued in service to the present day. It has been modified from time to time, but it has been found capable of fulfilling the requirements of engineers as regards its performance in high output engines, and of foundry production engineers and metallurgists, in its suitability for the production of sand-, gravity- and pressure-die cast parts in high quantity production. When manufacture of the particular engine, to which Mr. Devereux referred, was undertaken in America, during the war, attempts to replace this alloy with one of more modern development were unsuccessful, and this alloy has since been used exclusively in America for this job.

Continuing, he referred to the lines of development which lie ahead in the field of aluminium casting alloys. In the first place, the greatest demand will be for alloys which will enable cheap reliable castings to be made. This will mean that alloys will be simple in constitution, have good casting properties, especially in respect of gravity- and pressure-die casting, will not require elaborate and expensive heat-treatments to develop mechanical properties adequate for the majority of purposes to which they may be applied, good resistance to corrosion and ability to take surface finishes, particularly coloured anodic coatings, which will suit them to a wide variety of surroundings.

On the one hand we shall develop alloys for special purposes. We shall, no doubt, be asked for, and shall achieve, increasingly high mechanical properties in heat-treated materials. We shall continue the work of improving the properties of alloys at elevated temperatures in order to adapt them to the increased demands of all kinds of engines, including those of the internal combustion turbine, which is now developing so rapidly.

On the other hand, we shall see the development of low- and medium-strength alloys, particularly for pressure-die casting purposes, for a wide variety of applications where low cost will be the over-riding considerations and durability will be far more important than mechanical strength.

The field of magnesium casting alloys will probably provide the most important developments. It is only within the past two years that we have come to understand the corrosion behaviour of magnesium. The important work of Dr. Hanawalt on the effect of very small impurities on the corrosion resistance of magnesium has coincided with, and perhaps was partly suggested by, the new processes for producing magnesium which have come into operation since the outbreak of war. Some of these processes, in particular the thermal reduction processes, yield a product remarkably low in those impurities which have now been found to be so severely detrimental to the corrosion resistance of magnesium and its alloys.

With this new high-purity magnesium now available in commercial quantities at commercial prices, we are presented with great opportiunties for the use of magnesium castings in applications where, hitherto, they would have suffered severe corrosion. At the same time we have evolved protective processes which offer greatly improved resistance to corrosion.

The present control price of magnesium ingot 1/6 per pound, yet the high-purity metal from Canada could be sold in this country, at the present time, for 1/2 per pound, and it is hoped to be able soon to sell it at an even lower price.

Research and Development

In the field of scientific research and development the importance of research by individual firms was stressed, as well as of co-operative research by associations. In the future, no firm, least of all one in a highly technical industry like the founding of light alloys, will be able to carry on without conducting its own research into its own particular problems. No amount of collective research will make up the leeway between the firm which applies research methods to its production and that which depends on information handed out to all in the industry.

Conclusions

The whole of British Industry has a heavy responsibility in the critical years which lie ahead of our country. Upon its ability to maintain itself at the highest pitch of efficiency will depend our place in international trade and our ability to import the food and raw materials we require, the fulfilment of our plans for a better Britain and the way of life of our people. In this tremendous task the light-alloy foundry has an important play to play. In my opinion it is in fine condition to undertake its responsibilities. Great new efficient foundries are available, teams of managers, technicians and craftsmen of proven ability are eagerly awaiting the signal to start on the new tasks which lie ahead. Light alloy castings stand in high esteem among engineers, who, provided their confidence is maintained, will find increasing uses for these castings.

Research on alloys, on methods of moulding, pattern making and on die sinking; on casting methods and equipment and on the many other aspects of foundry technique must blaze the trail which must be closely followed by the application of new knowledge to production practices. We must not lose the spirit of adventure and the ability to take chances which we have learnt again during the war and in which we were sadly lacking in the years of "safety first" which preceded it.

High among our plans must rank the means of attracting to our industry young men and women of ability to ensure a continuous supply of new blood at all levels. Conditions of work and pay must bear comparison with those of other industries.

Bibliography on Cold-Heading and Cold Forging

VERY little data on cold-heading is in existence and that which is available appears in periodical literature only. There does not appear to be any comprehensive work on the subject, in fact, as far as can be ascertained, no up-to-date text book in English has been published devoting special attention to the subject. There is need for a fuller knowledge of this important branch of engineering and, in order to bring it to the notice of interested readers, an endeavour has been made to compile a list of references on the subject covering the period 1933–1943.

The arrangement is first of all by year: then author, title of article; followed by journal reference, in the order—name, part, date and pagination.

Arganbright (A. B.). "Basic open-hearth carbon steel for cold heading—'Ingot to wire.'" American Soc. for Steel Treating—Advance Paper No. 19 meeting. Oct. 2–6, 1933, p. 7.

Graham (H. W.). "Requirements of steel for cold heading." *Metal Progress* **24**. Aug., 1933, pp. 22–6.

Pulsifer (H. B.). "Materials for cold heading." *Metal Progress* **23**. Mar., 1933, pp. 12–7.

Harvey (C. L.). "Effect of cold-working on physical properties of cold-headed bolts." *American Soc. Metals—Trans.* **22**. July, 1934, pp. 657–69, disc. 669–72.

Seymour (L. D.). "Cold-heading wire." *Wire and Wire Products* **10**. Sept., 1935, pp. 414–8.

Alexander (J. C.). "Cold-heading dies and die steels." *Heat-treating and Forging* **22**. Apr., 1936, pp. 171–3.

Harvey (C. L.). "Cold-heading problems." *Wire and Wire Products* **12**. July, 1937, pp. 359–61 and 383.

Jones (R. H.). "Recent progress in cold-heading machinery." *Wire and Wire Products* **12**. Sept., 1937, pp. 479, 482–3, 501.

Rayn (A. R.). "Developments in cold heading at General Electric." *Wire and Wire Products* **12**. Oct., 1937, pp. 601–4, 606–7.

Smith (R. H.). "Cold-heading—bolts, rivets, nails." *American Soc. Metals—Trans.* **25**. Mar., 1937, pp. 84–96, 97. "Cold forging." *Steel* **102**. May 16, 1938, pp. 44–47.

Hubbard (G.). "How would upset parts fit your picture" *Machine Design* **10**. Jan., 1938, pp. 43–5.

Oliver (F. J.). "Appraisal of cold-heating practice." *Iron Age* **141**. June 9, 1938, pp. 28–32; June 23, pp. 24–8; **142**, July 7, 1938, pp. 42–5.

Palmer (F. R.). "Getting most out of cold-heading dies." *Iron Age* **142**. July 21, 1938, pp. 30–2.

Chase (H.). "Cold-forged parts." *Product Engineering* **10**. Jan., 1939, pp. 2–4.

Hopkins (H. L.). "Some characteristics of wire and their relation to cold heading." *Wire and Wire Products* **14**, Jan., 1939, pp. 36 and 38–9; **13**, Oct., 1938, pp. 521–3, 526, 528 and 595.

Christin (R. E.) and Freter (R. J.). "Cold-forge fabrication." *Tool Engineer* **9**. Sept., 1940, pp. 46–8 and 50.

Jameson (A. S.). "Hot and cold heading." *Metal Progress* **38**. Nov., 1940, pp. 691–7.

Reeder (A. M.). "Brief résumé of special cold-heading wire requirements." *Wire and Wire Products* **15**. Feb., 1940, pp. 99–102 and 130.

Chase (H.). "Cold header or screw machine." *Iron Age* **148**. Dec. 4, 1941, pp. 62–7. "Cold-heading die quenching fixture." *Iron Age* **150**. July 16, 1942, p. 53.

Jameson (A. S.). "Bolts—material, heat-treatment and physical properties." *Iron Age* **149**. May 7, 1942, pp. 57–60 and May 14, pp. 52–7 and 129.

Chase (H.). "Design of cold-headed parts." *Metals and Alloys* **18**. Aug., 1943, pp. 292–7.

Chase (H.). "Heat-treating cold-headed products." *Steel* **110**. Mar. 2, 1942, p. 72 and p. 74.

Estimation of Wetting Agent Concentration

By A. B. Ashton, M.Sc. and N. J. Stead

Surface active materials, or wetting agents have in recent years found increasing application in electroplating and allied processes. Sometimes the results have been disappointing due largely to the lack of closely controlled conditions. Operating on a commercial scale the difficulty has been to make due allowance for wastage ; for this purpose a reliable routine test is necessary and a method of estimating the amount of wetting agent present in the solution is described which is both rapid and simple.

THE use of additions of wetting agents in wet chemical processes is becoming common. So long as they are used under closely controlled laboratory conditions it is easy to control their concentration, but when used on a commercial scale for continuous operation, wastage of solution (due to drag-out, leakages, spray, overflows, discards, etc.) entails loss of the wetting agent. This loss has, of course, to be restored. Furthermore, there is in some cases the possibility of slow chemical decomposition of the compound by other materials in solution with it, and again loss due to this cause will have to be made good. These considerations show the need for a suitable method of estimating the amount of wetting agent present in a sample of solution. To be commercially valuable, the method should be rapid and simple, so that daily tests may be made a routine matter.

The apparatus and method described below were developed for the control of wetting agent concentration in a system containing some 12,000 gallons of sulphuric acid-copper sulphate solution. The system consists of a group of tanks for the pickling of oxidised copper and an electrolytic plant for the recovery of the dissolved copper, the solution being circulated continuously round the two sections of the system. The method allows the required concentration to be determined in duplicate in some 15 minutes, including the necessary volumetric copper estimation.

Chemical methods of analysis did not seem to be available. The marked effect of wetting agents in lowering surface tensions of liquids, however, suggested that a purely physical method might prove serviceable. Thus, if the relationship between wetting agent concentration and surface tension be suitable, measurement of the latter will give an indication of the former. The present method is based on this principle. Surface tensions are not directly measured but advantage is taken of the phenomenon of capillarity, which is directly due to surface tension.

The height reached by a liquid in a capillary tube as a result of surface tension is given by the formula :—

$$H = \frac{2T \cos \alpha}{rdg}$$

where H = capillary height in cms., T = surface tension in dynes per cm., α = angle of contact, r = radius of capillary in cms., d = density of liquid, g = gravitational constant.

Thus, if α, r, d and g are constant, capillary height is directly proportional to surface tension and, for the present purpose, may be taken as representing it. Thus the essence of the present method is the accurate measurement of the capillary height reached by a solution, followed by reference to a graph relating capillary height with wetting agent concentration, the graph being made by the use of a series of known standard solutions.

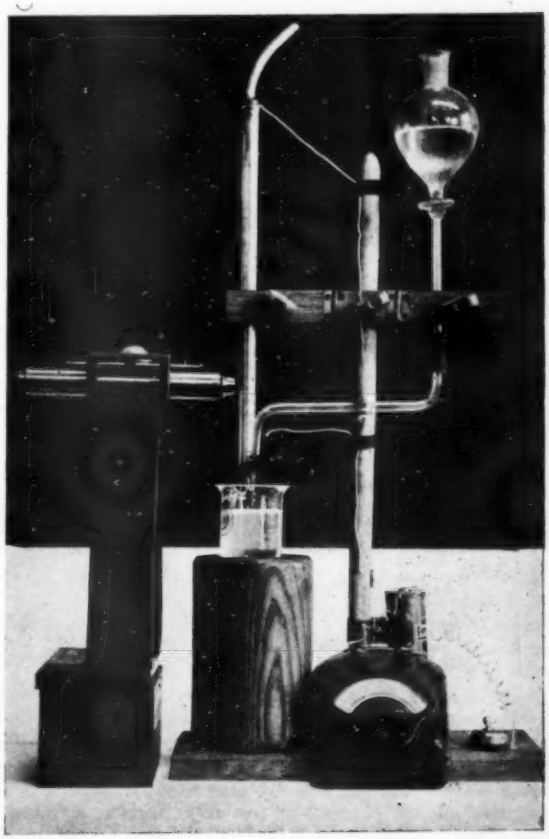


Fig. 1.—The apparatus used in the experimental work.

The wetting agent used in the work here described is a proprietary one marketed under the name of "Teepol X."

Preliminary experiments showed that three principal difficulties had to be overcome : firstly that of reading the capillary height sufficiently accurately, since the depressions obtained are small ; secondly, of adjusting

the solution level outside the capillary to exactly the same point for each determination; and, thirdly, in cleaning the capillary thoroughly but rapidly after each test. The apparatus as shown in Figs. 1 and 2 was accordingly designed to surmount these three difficulties and was used successfully as described below.

Construction of Apparatus

The tap and ungraduated portion at the bottom of a 25 cc. burette were removed and a capillary of 0·4 mm. bore and 20 cm. long, widening out to a tube of 6 mm. bore and 5 cm. long, was mounted in it by passing the wide part through a one-hole cork at the top of the burette tube and allowing the capillary to project about 2 cm. through a slot in a cork at the bottom of the tube.

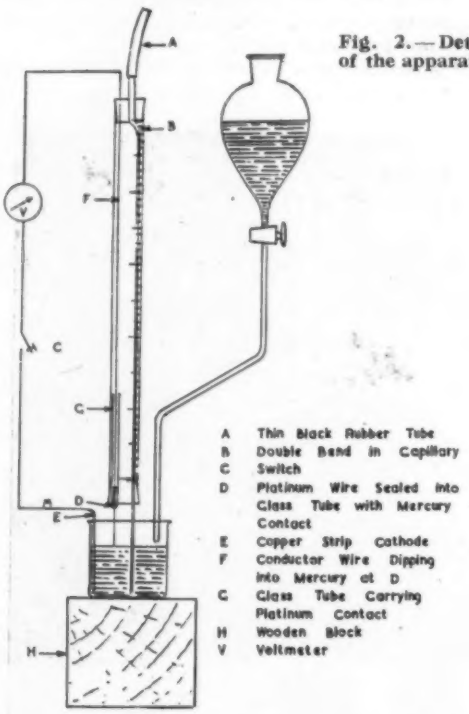


Fig. 2.—Details of the apparatus.

A double bend was made in the capillary at its upper end so that it was in contact with the inner wall of the burette on the same side as the scale over its whole length, thus avoiding errors in reading due to parallax. The capillary height could then be read to the first decimal place from the burette scale and to the second-place from the eyepiece scale of a microscope mounted horizontally in a convenient position, the burette thus providing a quite arbitrary but very convenient scale of units.

The burette tube was mounted vertically in an adjustable stand so that the projecting lower end of the capillary reached approximately half the depth of a 100 cc. beaker, raised above bench level on a movable wooden block to facilitate changing the beaker.

It then only remained to provide a means of adjusting the liquid level in the 100 cc. beaker to an exactly fixed height for each test and this was accomplished by inserting a glass tube through the lower cork in the burette tube carrying a short length of platinum wire sealed through the glass and projecting downwards. Contact was made with the wire internally by a globule of mercury. A thin copper wire lead dipping into the mercury was brought out through the top of the burette tube and connected through a switch and voltmeter to the positive terminal of a 3-volt dry cell. The negative terminal was connected to a thin copper strip about 5 cm. × 1 cm. dipping into the 100 cc. beaker.

A 150 cc. separating funnel was then arranged as shown, so that the liquor under test could be run from it into the 100 cc. beaker; at the moment when the liquid reached the platinum contact the electrical circuit was completed via the liquor and the strip cathode and a voltmeter deflection was observed. Thus, by running in the liquor drop by drop a very accurate reproduction of liquor level could be maintained in successive tests. A 2·5-volt bulb was at first used instead of the voltmeter but the circuit resistance was too high to allow it to illuminate and it was abandoned in favour of the voltmeter, which is operated by potential rather than current.

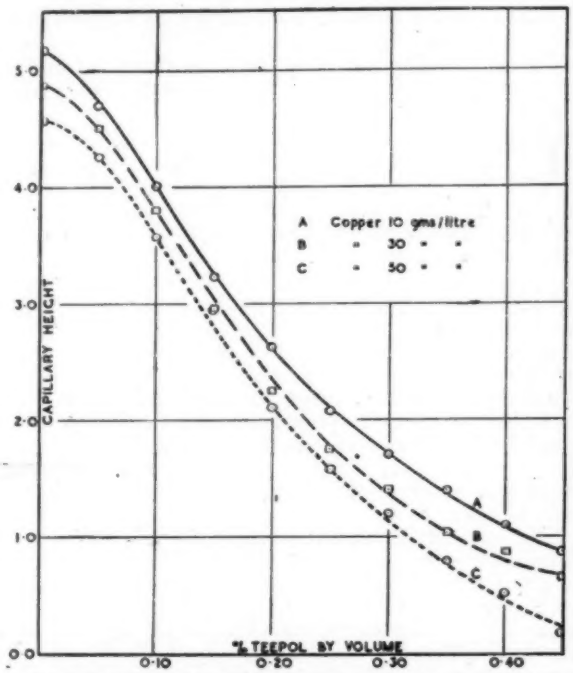


Fig. 3.—Standard Curves for copper concentrations of 10, 30 and 50 g/l.

Procedure

It was found to be essential to adhere to a strictly uniform procedure in using the apparatus, and the method as detailed below was devised.

100 cc. of the liquor to be tested are poured into the separating funnel, the dry cell switched into circuit and the liquid allowed to run slowly into the 100 cc. beaker, the flow being adjusted to a drip when near to the required level, and stopped immediately a voltmeter deflection is observed.

A short length of thin black rubber tubing is then attached to the wide portion at the top of the capillary and manipulated so that liquor is drawn in to fill

the capillary completely. The liquor is then expelled and drawn to the top of the capillary once more, great care being taken to see that no air bubbles are obtained either in the capillary itself or at the point where it dips into the liquor. The rubber tubing is now gently squeezed until the liquor in the capillary falls to the surface level in the beaker and then released so that the liquor rises due to the surface tensional force alone. It is observed through the microscope and, when a constant height is attained, the burette scale and eyepiece scale are read and the composite figure recorded.

The liquid is then expelled from the capillary, the 100 cc. beaker removed and replaced by another containing water. Water is drawn into the capillary and expelled, a beaker containing chromic acid then being substituted for the water, and the procedure repeated. The capillary is again washed with water, then with concentrated sulphuric acid and finally with water.

The capillary is then clean for the next test but the procedure for cleaning must be repeated between each determination since absolute cleanliness of the capillary is essential.

Calibration of the Apparatus

The calibration was carried out in two stages, the first step being to observe fluctuations in capillary height due to variables other than "Teepol," namely, temperature, acid concentration and copper concentration; the second being to maintain each of these factors at a constant amount and measure the capillary height for various "Teepol" contents. The results obtained are given in Tables I, II, III and IV.

TABLE I.

RELATION BETWEEN CAPILLARY HEIGHT AND TEMPERATURE OVER A NORMAL WORKING RANGE.

Temperature ..... 17·5–29·5° C.
Copper Concentration ..... 16 g/l.
Acid Concentration ..... 10 g/l.
"Teepol" content ..... Nil.

Temperature, °C.	Capillary Height
17·5	4·90
29·5	4·90

NOTE.—In this and subsequent tables units for the capillary height are the burette graduations and, therefore, arbitrary.

TABLE II.

RELATION BETWEEN THE CAPILLARY HEIGHT AND THE ACIDITY OVER A NORMAL WORKING RANGE.

Temperature ..... 18° C.
Copper Concentration ..... 25 g/l.
Acid Concentration ..... 10–50 g/l.
"Teepol" content ..... Nil.

Acid Concentration	Capillary Height
10 g/l.	4·91
20 g/l.	4·92
30 g/l.	4·91
40 g/l.	4·90
50 g/l.	4·90

It will be seen from the Tables that the capillary height was not seriously affected by temperature or acid variations (Tables I and II) but that it decreased as the copper concentration increased (Table III).* For the second stage, therefore, three solutions were used with copper concentrations of 10 g/l, 30 g/l and 50 g/l, the acidity being constant at 25 g/l. Results shown in

* This phenomenon is due to the fact that, over the ranges of concentration tested, $CuSO_4$ has a greater effect on specific gravity of liquor than H_2SO_4. Capillary height is inversely proportional to the specific gravity of the liquid.

TABLE III.

RELATION BETWEEN THE CAPILLARY HEIGHT AND THE COPPER CONCENTRATION.

Temperature ..... 18° C.
Copper Concentration ..... 10–50 g/l.
Acid Concentration ..... 10 g/l.
"Teepol" content ..... Nil.

Copper Concentration	Capillary Height
10 g/l.	5·21
20 g/l.	5·03
30 g/l.	4·88
40 g/l.	4·72
50 g/l.	4·58

TABLE IV.

RELATION BETWEEN THE CAPILLARY HEIGHT AND THE "TEEPOL" CONTENT.

Temperature ..... 18° C.
Copper Concentration ..... 10, 30, 50 g/l.
Acid Concentration ..... 25 g/l.
"Teepol" Content ..... 0–0·45%

"TEEPOL" ADDED %	COPPER 10 g/l. Capillary Height.	COPPER 30 g/l. Capillary Height.	COPPER 50 g/l. Capillary Height.
0·0	5·18	4·88	4·57
0·05	4·70	4·50	4·23
0·10	4·00	3·80	3·58
0·15	3·23	2·94	2·96
0·20	2·62	2·27	2·09
0·25	2·07	1·75	1·57
0·30	1·70	1·40	1·20
0·35	1·40	1·03	0·80
0·40	1·10	0·88	0·52
0·45	0·87	0·65	0·17

TABLE V.

RELATION BETWEEN ACTUAL AND OBSERVED "TEEPOL" CONCENTRATIONS.

Temperature ..... 18° C.
Copper Concentration ..... 30 g/l.
Acid Concentration ..... 25 g·l.
"Teepol" ..... 0·05–0·30%

"Teepol" Added, %	Capillary Height	"Teepol" Found, %
0·05	4·51	0·05
0·10	3·79	0·10
0·15	3·28	0·13
0·20	2·25	0·20
0·25	1·68	0·26
0·30	1·39	0·30

Table IV were obtained. The results shown in Table IV were plotted and are shown in Fig. 3, which gives three curves, one for each of the copper concentrations, representing the capillary heights obtained when various amounts of "Teepol" are present. From these curves the effect of added "Teepol" on the pickling liquor can clearly be seen. The "Teepol" content may be determined in any given pickling liquor by determining its copper content and the capillary height obtained, after which the "Teepol" percentage is read off from the appropriate graph. The "Teepol" concentration in liquors of copper contents other than the three for which curves are given can be determined with sufficient accuracy for commercial purposes by interpolation.

Typical Results Obtained

Table V shows "Teepol" contents determined by the above method on acid samples to which exactly measured amounts of "Teepol" had been added. The agreement between actual and observed values is seen to be quite close enough for commercial purposes. It should be noted, however, that the accuracy of the method appears to decrease with amounts of "Teepol" greater than 0·30%; since at above that figure the rate of change of capillary height with "Teepol" concentration becomes small.

If it be desired to ascertain actual values for surface tension, this can be done with moderate accuracy by a little calculation. The units used as described above are purely arbitrary ones derived from the burette scale. If, however, the capillary height of water (of which the surface tension is known) be determined in the apparatus these units can be converted into real values of surface tension. It was found that the surface tension of a sample of pickling acid was lowered from 69·1 to 41·9 dynes per cm. by the addition of 0·2% "Teepol."

The authors' thanks are due to the management of the London Electric Wire Co. and Smiths, Ltd., for permission to publish this information.

Hardenability of Steel

Considerable attention was given to hardenability of steels at a recent meeting of the American Society of Metals. Such aspects as isothermal transformation and end-quench hardenability of some N.E. steels, further developments of the end-quench hardenability test, a hardenability test for low-carbon and shallow hardening steels, the effect of hardenability and the air hardenability of steels, were discussed. This composite article summarises the views expressed and gives the results of investigations.

ISOTHERMAL transformation diagrams, determined by the metallographic method, are given by R. L. Rickett, J. G. Cutton, C. B. Bernhart and J. R. Millikin,[1] for steels of the 8600 to 8900 series, containing 0·4 to 0·6% nickel, 0·4 to 0·6% chromium, 0·75 to 1·0% manganese and varying amounts of molybdenum; and for steels of the 9400 to 9600 series, introduced to economise still further in nickel, chromium and molybdenum and containing 0·4 to 0·6% silicon. As some of the steels investigated have since been withdrawn from the N.E. Steel Schedule and replaced by steels with a silicon range of from 0·2 to 0·35%, a list of such current steels is given for comparison with the N.E. steels tested. End-quench curves were also determined for some of the steels.

The isothermal diagrams obtained were of the same type as those obtained for many S.A.E. alloy steels and show a bulge in the vicinity of 540° C. In the same range of temperature a "secondary nose" occurred, representing the beginning of formation of ferrite carbide aggregate and indicated the possibility of obtaining intermediate-temperature transformation products in such steels by continuous cooling within a particular range. A similar resemblance was also obtained between the microstructure of isothermally transformed N.E. steels and those of Standard S.A.E. alloy steels. Above 540° C. proeutectoid ferrite and ferrite was formed; between 480° and 540° C. there was a formation of accicular ferrite containing carbide particles, and at still lower temperatures an accicular iron-carbide aggregate (bainite) was formed. The end-quench curves indicated the relatively hardenability which could be predicted from the isothermal transformation diagrams and gave a more precise measure of the hardenability of a steel. Both methods of examination are considered as useful in supplementing one another.

Development of the End-Quench Hardenability Test

In dealing with further developments of the end-quench hardenability test, C. R. Wilks, E. Cook and H. S. Avery[2] discuss the limitations of the standard Jominy end-quench hardenability test and propose the use of a hardenability test bar selectively quenched from both ends, in order to provide correlations with the thermal histories of quenched plates that correspond more closely than is possible with the standard single-end-quench specimen. In carrying out the tests, double ended cupped test specimens, 1¼ in. to 8 in. long and 1 in. to 2 in. diameter, were quenched in a modified fixture, using a water flow of 194 ft./min. and a water temperature of 750° ± 3° F.

Cooling curves were obtained at the centre ($\frac{1}{32}$ in. below the surface) of double end quench specimens of a manganese-chromium-molybdenum steel quenched from 900° C.; a comparison was made of centre and surface cooling curves midway between ends of double end quench specimens of the same steel, and the effect of diameter on the cooling curve of a 2⅜ in. double end quench bar was also determined. The effect of composition upon the centre cooling curve for a 7⅞ in. double end quench bar was determined by quenching a carbon steel, an alloy steel and a grey cast iron from 900° C., and centre cooling curves for 2⅞ in. and 5⅞ in. specimens of a 0·47% carbon steel double end quenched from 900° C. were also obtained. From these curves it was found that for those steels where pearlitic reaction may be avoided, but which transform partially or wholly under the same cooling conditions to acicular structures of the bainite type, the test could be useful in predicting behaviour at the section centre, since the mechanical properties at the centre of symmetrical double end quench specimens might be determined. Centre cooling curves for sand blasted ⅞ in., 1¼ in., and 1⅞ in. plates, quenched in agitated water were compared with those for selected double end quench specimens and a useful correlation of specimen thickness against plate thickness, based on these curves, was found, and the utility of the double end quench specimens for determining mechanical properties after typical plate heat-treatment indicated. Special applications where the double end quench technique might be superior to the single end quench practice are suggested and several variables discussed.

O. W. MacMullan[3] discusses the end-quench test method developed by Jominy and Boegehold, which has been widely accepted as the standard method for determining hardenability of steels which develop a considerable degree of depth hardness and have higher carbon content than is usually found in carburising

1 *Amer. Soc. for Metals.* Preprint 5, pp. 1–22.
2 *Amer. Soc. for Metals.* Preprint 6, pp. 1–20.
3 *Amer. Soc. for Metals.* Preprint 7, pp. 1–19.

TABLE I—HARDNESS RESULTS FROM THE FOUR TYPES OF TEST SPECIMENS. SHALLOW HARDENING AND CARBURISING SERIES.

No.	Type	Specimen	Rockwell "C" at distance shown $\frac{1}{16}$in.																						
			1	2	3	4	5	6	7	8	9	10	11	12	13	14	15	16	17	18	19	20	22	24	28	32	40
1	1022	Standard	38	25	17	13	12	11	8·5	7	—	6	—	6	—	4·5	—	4	—	—	—	2	—	1	0	—	—
		45° C.	39	34	26·5	19	16	13·5	12	10	9	8	7	6	—	5·5	—	4·5	—	3	—	2·5	—	—	—	—	—
		Type L	43·5	42	41	38	36·5	35	33	30·5	28	25	23	18	16	14	12	11	—	9	—	7·5	7	5·5	3·5	2	1
		26½° R.	42	39	36	35	33	29	26	23	21	18	16·5	15	13	12	11	10	—	8·5	—	8	—	7	5·5	5	—
4	1045	Standard	57	53	37	31	29	28	27	26	—	25	—	24	—	23	—	22	—	—	—	19	—	17	16	15	14
		45° C.	61	39·5	58	55	48	39	32	29	27·5	—	26	—	25	24	—	—	22	—	21	19·5	—	—	—	—	—
		Type L.	63	62	—	62	—	61	60	59	58	56	54	50	48	44	39	34	31	30	—	29	27·5	26	23	21	17
		26½° R.	—	61	—	61	—	60	59	58	57	55	52	47	40	33	31	29	—	27	—	26	25	24	23	22	—
7	1080	Standard	67	66	58	45	44	—	—	43	—	42	—	41	—	39·5	—	38	—	—	—	36	—	33	31	29	28
		45° C.	66	66	—	66	65·5	64·5	59	46	44	—	43	—	42	—	41	40	39	38	—	—	—	—	—	—	—
		Type L.	67	—	—	—	—	67	—	—	—	66	—	66	—	65	54	56	45	44	—	43	—	41	38	35	31
		26½° R.	—	65	66	—	—	66	—	66	—	65	—	64	63	60	54	46·5	—	44·5	—	43	—	42	40	38	—
10	4815	Standard	42	42	39	34	30	26·5	25	24·5	—	22	—	21	—	20	—	19	—	—	—	18·5	—	17	16	15	15
		45° C.	42	43	43	41	39	37·5	36	35	31·5	29	28·5	—	27	26	—	26	—	26	—	—	—	—	—	—	—
		Type L.	44	45	—	45	—	45	—	44	—	43	—	43	—	42	40	39	36	33	30	28	2·5	22	19	18	16
		26½° R.	—	43	43·5	44	—	43·5	43	42	41·5	40·5	39	38	37	36	34	32	—	30·5	—	28·5	—	27	25	24	—
15	8620	Standard	47	45	39·5	33	29	27	26	24	—	23	—	21	—	19	—	17	—	—	—	16	—	15	15	14	10
		45° C.	48	47	47	46	45	43	41	38	35	33	32	—	31	—	31	30	30	29	28	27	—	—	—	—	—
		Type L.	48	—	48	—	—	48	—	47	—	46	45	44	43	41	38	35	32	30·5	29	27	24	22	18	16	12
		26½° R.	—	47	47·5	—	48	—	47·5	47	46	45	44	42	41	39	37·5	36	34·5	33·5	32·5	31·5	30·5	29	27	25	—

steels and which may be regarded as suitable for use on steels containing more than 0·25 to 0·30% carbon. On the other hand, there is a lack of general acceptance of this or any other method for the carburising grades and for the shallow hardening steels. A Jominy L type specimen, solid cylinders, wedge and cone type specimens, have been used, but are found to possess certain disadvantages. A new method is, therefore, proposed for determining the hardenability of such steels and is claimed to combine the desirable features of the end-quench and wedge types of test and to eliminate certain objections to both methods.

In the new method the specimens are tapered or wedge-shaped and are jet quenched on one face only of the wedge in the same manner as used for the end quench. Sectioning after hardening is unnecessary and hardness readings, instead of being made on a surface perpendicular to the quenched face, are taken along the other face of the wedge, forming an angle with the quenched face. By changing the angle the fully hardened and transition zones can be extended along the oblique face on which readings are taken to fit any degree of hardenability within limits of reasonable specimen size.

Specimens used were of two designs. Those formed by cutting off the ends of a 1 in. round bar at an angle of 45° and those cut from a square or rectangular bar to form a more acute angle, 26·5°. To use the standard Jominy equipment for quenching, specimens were attached by means of a small rod to the collar end of shortened Jominy bars. Tests were carried out on a number of plain and alloy steels, including nickel-molybdenum carburising type and low-alloy N.E. steels of the 8720 and 9420 series. Hardness results from four types of test specimens with shallow hardening and carburising steels are given in Table I.

The conclusions drawn from these tests were that hardened zones were widened along the surface and that a satisfactory range of hardness was obtained in alloy carburising steels from the 1 in. round specimens cut at 45°, tangent 1. Wedge specimens increased the length of the hardened zones and the total spread of hardness. The mass of hot metal behind the quenched face influenced the cooling rate, consequently readings on tapered specimens were not directly convertible to straight bar results at the same distance from the quenched surface. Reproducibility of results by end quench procedure, in spite of a number of variables, was confirmed with specimens of various sizes and shapes.

A Hardenability Test for Low Carbon Steels

E. S. Rowland, J. Welchner, R. G. Hill and J. J. Ross, have determined the effect of carbon content on the hardenability of alloy steels. The investigation was

4 *Amer. Soc. for Metals.* Preprint 16, pp. 1–27.

TABLE II.—COMPOSITION, TEMPERATURE OF TREATMENT, AND HARDNESS ATTAINABLE IN VARIOUS AIR HARDENING STEELS AS A FUNCTION OF A/V.

Composition.							Hardening Temperature °C.	Minimum A/V for Minimum Rockwell "C" of						
C	Mn	Si	Cr	Ni	W	Mo		64	62	60	55	50	45	40
0·93	0·62	0·24	3·68	0·20	—	0·93	920	3·0	1·4	1·1	<1·0	—	—	—
							950	3·0	1·0	<1·0	<1·0	—	—	—
							980	Nil	Nil	<1·0/1·4	<1·0	—	—	—
							1100	Nil	Nil	Nil	Nil	<1·0	—	—
0·39	0·40	0·99	5·10	0·09	1·22	—	980/1035	Nil	Nil	Nil	<1·0	—	—	—
2·16	0·40	0·40	12·00	0·48	—	0·087	925	Nil	3·0	2·2	1·7	1·6	1·3	1·1
							950	3·8	2·7	2·4	2·0	1·7	1·4	1·1
							980	2·6	2·1	2·0	1·7	1·5	1·1	1·0
							1010	Nil	2·3/3·3	2·0	1·6	1·2	1·1	1·0
1·50	0·35	0·36	12·09	0·35	—	0·83	1010	1·6	<1·0	—	—	—	—	—
1·13	0·42	0·34	16·70	0·36	0·02	—	1023	Nil	Nil	1·41	<1·0	—	—	—
0·35	0·41	0·40	12·90	0·50	—	0·26	1010	Nil	Nil	Nil	1·4	<1·0	—	—
0·104	0·36	0·26	1·54	3·67	—	—	840	Nil	1·3	<1·0	—	—	—	—
0·57	0·56	0·71	7·45	0·22	7·77	0·15	1035	Nil	1·0	<1·0	—	—	—	—
0·71	0·47	1·17	3·04	0·32	—	5·63	1035	2	<1·0	—	—	—	—	—
0·26	0·49	0·17	1·24	4·28	—	—	830	Nil	Nil	Nil	Nil	Nil	Nil	<1·0
0·69	2·09	0·35	1·08	0·15	—	1·57	—	—	—	—	—	—	—	—

conducted on a group of induction heats, varying in carbon content in approximately 0·20% increments from 0·2% to over 1% in each of two types of steels: S.A.E. 52100 (1·5% chromium approximately) and S.A.E. 4620 (1·76% nickel, 0·26% molybdenum approximately). End quench hardenability tests were run on bars from these heats from two prior structures, annealed and normalised, and the effects of variation in time at quenching temperatures up to 4 hours and quenching temperatures up to 925° C. Hardenability results are given as a function of carbon content under all the conditions investigated.

From the results obtained it was concluded that at normal hardening temperatures, the hardenability of a given alloy series increased with the carbon content to a maximum value corresponding to the position of the A_{-} line under the conditions of testing imposed and then decreased with further increase in carbon content and this decrease was due to the nucleating effect of the excess carbide on the decomposition of austenite on quenching. The similarity in hardenability behaviour between the two alloy series was so marked that it was assumed that the change in direction of the hardening behaviour under the influence of change in time, temperature and prior structure could be predicted with some confidence from a knowledge of carbon content alone, and that the type and amount of alloying elements controlled the extent of the hardenability change.

In commercial hypoeutectoid steels, annealed prior structures increased in hardenability only slowly with increase in time at normal quenching temperatures, while normalised prior structures assumed maximum hardenability at very short time intervals and were substantially insensitive to quenching temperature variations between the Ac_3 and the coarsening temperature, while the former increased continuously in hardenability with increase in quenching temperature. With normal commercial soaking times and quenching temperatures, hypereutectoid steels produced maximum hardenability from annealed prior structures rather than from normalised prior structures and this was considered due to and controlled by the number and distribution of those excess carbides which act as nuclei for austenite decomposition. In the eutectoid range the hardenability change was enormous, with variations in condition, but, in general, followed the same laws as in lower carbon ranges. The normalised prior structure produced higher hardenability than the annealed under all reasonable conditions of time and quenching temperature.

The Air Hardenability of Steels

In the last paper[5] of the series, C. B. Post, M. C. Fetzer and W. H. Fenstermacher discuss experimental data obtained on the air hardenability of steels and describe an air hardenability test and a method of assessing hardenability in a manner applicable for the use of tool makers and others in the production and selection of air-hardening steels. The test developed consisted of air cooling a 1 in. round bar in such a manner that one end was allowed to cool freely while the cooling of the other end was retarded to a rate equivalent to that of a 6 in. round × 6 in. long cylinder. A determination of the hardness gradient along the 1 in. round bar enabled the Rockwell hardness to be stated as a function of the cooling velocities. The parameter found to give the closest agreement between hardness predicted from the air hardenability test and that obtained by actual heat-treatment of various size sections was the cooling rate (degrees F. per min.) past 1000° F. (540° C.).

The most obvious application of the test is the prediction of the hardness to be expected of a given air hardening steel when air treated in various sized pieces. The application of these data is simplified by the fact that cooling rates in the vicinity of 540° C. are equal throughout the mass during air hardening, except for extremely irregular shapes having fins, pins, etc., protruding from the larger mass, which means that air hardening leads to the same hardness throughout the mass from centre to surface. Given, therefore, the centre cooling rate of a regularly shaped mass and the air hardenability of a particular air hardening steel, it is easy to predict the hardness which will be obtained in the mass when using this air hardened steel. The simplicity of this application of air hardenability data depends in a large measure on the effect of (*a*) shape of piece (round, rectangular, long or short), and (*b*) treating temperature on the cooling rate of a regularly shaped mass of steel.

Data showing that the air cooling rate (past 540° C.) of regularly shaped masses is a function of the surface/volume ratio (A/V) only and that this cooling rate for a given A/V appears to be independent of treatment, were obtained by investigations on 32% nickel steel and three air hardening steels. Air cooling rates past 540° C., ranging from three of a 1 in. round × 8 in. long (A/V = 4·25) to a 6 in. round × 6 in. long specimen (A/V = 1) were obtained simultaneously by means of the air hardenability test described, and permitted the hardness, developed by an air hardening steel to be determined as a function of the cooling rate past 540° C. The air hardenability of various types of steel was also determined, Table II, and shows that variations in heat-treatment and soaking time affected air hardenability, but that the rate of heating through the critical range had no effect.

5 *Amer. Soc. for Metals.* Preprint 17, pp. 1–22.

Mr. J. M. Osborne, A.M.I.C.E., has been appointed a director of Dorman, Long and Co. Ltd. Joining the Bridge Department of the company in 1926, Mr. Osborne became manager of Dorman Long's bridge-building activities five years later. In 1934 he went out to South Africa to report on the company's interests in the Union and in January, 1935, was appointed managing director of the Union Construction Company (Pty.) Limited, in which Dorman, Long were substantially interested.

Mr. J. M. Osborne has been closely concerned with the Union's industrial war effort. Chairman in 1939 of the Industrial Council of South Africa, a joint employers' and trades union body concerned with labour matters, Mr. Osborne was re-appointed chairman in 1940, on the nomination of the trades unions. In 1942 he was appointed a member of a Government Committee reporting on War contracts. Since its inception in 1943, Mr. Osborne has been a member of the South African Board of Supply. In September, last year, he was appointed a member of the South African Supply Mission which visited the U.S.A. and Great Britain to exchange views on the supply position in the post-war transition period. The Mission was also concerned with the co-ordination of supply arrangements following the end of the war in Europe.

Chromising Processes

A Survey of latest Russian and German Research

By D. W. Rudorff

During recent years considerable attention has been given to the chromising of surfaces on iron and steel parts with a view to counteracting corrosion. The process of diffusing chromium into the metallic surface depends for its success on the effective and uniform supply of chromium to the surface to be treated and also upon the maintenance of conditions favourable for diffusion in sufficient depth. The use of granular chromium or ferrochrome and of liquid chromium are briefly referred to in this review, but particular attention is directed to the use of the gaseous phase by the employment of chromous chloride which has made great progress in Russia and Germany.

AMONG the various methods recently developed for the production of a corrosion resistant surface stratum in iron and steel alloys, particular attention attaches to the chromising processes based upon the diffusion of chromium into the metallic surface. The two basic prerequisites to a successful performance of this operation will obviously be (*a*) an effective and uniform supply of chromium to the surface to be treated and (*b*) the maintenance of conditions favourable for a diffusion in sufficient depth in all parts of the surface to be protected. This is a matter of considerable importance in so far as corrosion resistance will be gravely impaired by the existence of any weak spots that may have their origin in local non-metallic inclusions in the surface material.

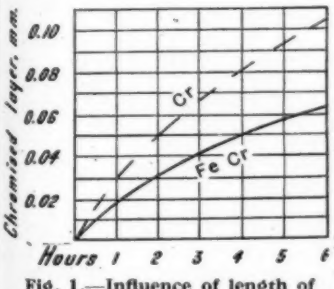


Fig. 1.—Influence of length of gas-phase treatment upon depth of chromised layer in 0.12% C. material.

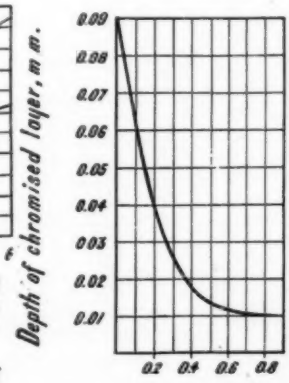


Fig. 2.—Influence of C control upon depth of chromised layer produced by gas-phase treatment for 4 hours at 980° C.

With regard to the first of these two problems, the choice must lie between an application of the chromium-bearing substance in either the solid, liquid, or gaseous phase, the application being due to physical contact or reaction with the surface metal.

The method of solid contact had been resorted to in early attempts, but had been found wanting, since the mere contacting of the surface with granular chromium-metal or ferrochrome at diffusion temperature does not produce diffusion zones of the required uniformity. It goes without saying that the application of chromium-metal in the liquid phase must remain restricted to a few exceptional cases in which the high temperature of the molten metal is no obstacle; while treating with chromium-metal in the gaseous phase is unsatisfactory because of the low vapour pressure of the metal.

With the employment of chromous chloride, however, use can be made of the reaction $CrCl_2 + Fe = FeCl_2 + Cr$, which makes it possible to produce chromium at the metal surface by contacting the latter with the chromous chloride. When applying this process it is, of course, necessary to ensure removal of the nascent $FeCl_2$ in order to keep the reaction going. Chromous chloride has a melting point of 820° C. and at *e.g.* 1,100° C. has a vapour pressure of approximately 25 mm. w.g. All gaseous chromising processes are based upon the utilisation of the vapour pressure.

Russian Research

In Germany, notable progress in the employment of the gaseous process has been due to the work of G. Becker, K. Daeves, and F. Steinberg [1] and the method has also been thoroughly investigated by the research workers of the Moscow Steel Institute.[2] The work of the Russian workers is of particular interest in so far as both the gas phase and the solid phase methods have been investigated on a comparative basis. One series of tests was concerned with the explanation of the influence of the process temperature upon the depth of case produced in the subjoined table.

INFLUENCE OF PROCESS TEMPERATURE UPON DEPTH OF THE CHROMISED LAYER IN 0·12% CARBON STEEL (2-HOUR TREATMENT) ACCORDING TO A. N. MINKEVITCH

	Process Temperature ° C.				
	950	980	1,100	1,300	1,380
Depth of chromised layer with gas-phase treatment, mm.	0·025	0·06	0·05	0·19	—
Depth of chromised layer with solid-phase treatment, mm.	0·013	—	0·03	0·1	0·16

In the solid-phase treatment the surface to be chromised was covered with a mixture of 55% Cr and 45% Al_2O_3, where the Al_2O_3 served to diminish the tendency of the Cr to stick to the surface of the steel.

The influence of the length of the chromising treatment upon the depth of the chromised layer was investigated at 980° C., the hydrogen consumption amounting to 40–50 litres per hour. In these tests chromising was carried out with the use of metallic chromium as well as with ferrochrome of 55·46% Cr and 0·25% C. The results of these tests are given in Fig. 1. Another series of tests was concerned with the influence of the flow rate of hydrogen upon the process, and it was found that with all other conditions remaining equal, the depth of

1 G. Becker, K. Daeves, F. Steinberg, *Stahl und Eisen.*, Vol. 61, 1941, p. 289. *Metallwirtsch*, Vol. 20, 1941, p. 217.
German Patent No. 706,112, April 10, 1941.
2 A. N. Minkevitch, *Stahl*, No. 5/6, 1945, p. 68.

the chromised layer increased with higher flow rates of hydrogen. Thus at 980° C. process temperature with a length of treatment of 4 hrs., the depth of the layer was measured as 0·03 mm., hydrogen being supplied at the rate of 20 litres per hr.; while with a flow rate of 30 litres per hr. a depth of 0·05 mm. was registered. With a flow rate of 50 litres per hr., the depth of the chromised layer increased to 0·08 mm. and a rate of 70 litres per hr. resulted in a depth of 0·11 mm.

By applying the gas-phase chromising process to steels of various carbon contents, it was found that increase in carbon content tends to decrease the thickness of the chromised layer produced. The materials investigated were: (1) Armco iron of 0·026% carbon, (2) a steel of 0·12% carbon, (3) a steel of 0·46% carbon, and (4) a steel of 0·81% carbon. The results obtained are charted in Fig. 2. The surface hardness of the chromised layers of steels (2) and (4) was determined in Vickers units. These test values are reproduced in the subjoined table.

Treatment	Steel No.	Vickers Hardness 1 kg. load
980° C., 4 hrs.	2	103–113
	4	876–958
1,150° C., 1 hr.	2	105–117
	4	876–1,008
1,150° C., 5 hrs.	4	1,103–1,219

Chromising treatment of the Armco iron for 6 hrs. at 980° C. produced a chromised layer of 0·1 mm. depth containing 25% Cr.

Gas-Phase Process

A considerable difficulty, which had to be overcome in the development of the gas-phase process, is the relatively high density of the gas which impedes the production of a homogeneous mixture of the gas with the type of carrier gas usually employed. But a recent German report by H. Bennek, W. Koch, and W. Tofaute[3] states that its authors have succeeded in finding a carrier gas (its composition is not revealed) which at sub-atmospheric pressures varying from 20–200 mm. w.g. forms a homogeneous mixture with the gaseous chlorides. For the commercial application of this method a special type of vacuum furnace was developed which operates at approximately 1,000° C. It is claimed that the chromised layers produced by this method are of excellent quality, exhibiting a smooth and uniform chromised layer even in spots difficult of access. The corrosion resistance of layers produced in this way is exemplified by the test results reproduced below :—

Steel	Loss in Weight in g/m², h, when immersed in			
	45% Nitric Acid, boiling	10% Acetic Acid, boiling	Mixture of 60% Vol H_2SO_4 20% Vol HNO_3 20% Vol H_2O at 60° C.	Artificial sea water at 20° C.
18·0% Cr, 8·0% Ni, plain	0·1	0·0	0·02	0·01
17·0% Cr, plain	0·3	0·02	0·07	0·01
0·1% C, 0·0% Ti, chromised ..	0·03	0·06	0·04	0·01
0·1% C, 1·5% Mn, 0·5% Ti, chromised	0·03	0·06	0·05	0·01
0·1% C, 3·5% Mn, chromised ..	—	—	0·05	0·01

According to these data, the corrosion resistance of the chromised materials against the attack of nitric acid is superior to that of the 18 Cr 8 Ni and of the 17 Cr materials, but it appears to be somewhat inferior in the case of exposure to acetic acid. In view of the shallow depth of the chromised layer, ranging from 0·1–0·2 mm. in practice, it is obvious that the protective effect of the chromised layer can be no more than temporary.

From the aspects of the physical chemistry of the various gas phase processes, the recent work by C. Wagner and V. Stein[4], and by C. Wagner[5] is highly important and must therefore be at least briefly mentioned. The outstanding problem treated by C. Wagner and V. Stein is the vapour pressure equilibrium conditions of Cr compounds, reference being made to the various patented gas-phase chromising processes as well as to the work of previous investigators. The vapour pressure equilibrium conditions of the system Cr-H_2-Cl for metallic Cr as well as for Cr-chlorides and the reaction Fe (Mixed crystal) + $CrCl_2$ (Gas) = Cr (Mixed crystal + $FeCl_2$ (gas) are treated in detail. The work of C. Wagner is devoted to an analysis of the diffusion process proper, special consideration being paid to the deactivation of the gas stream as it passes over the processed piece. Assuming that the decrease in the concentration of $CrCl_2$ in the carrier gas must not be allowed to exceed 25%, Wagner establishes an expression for the minimum flow rate at which the carrier gas must be made to pass over the surface of the treated piece.

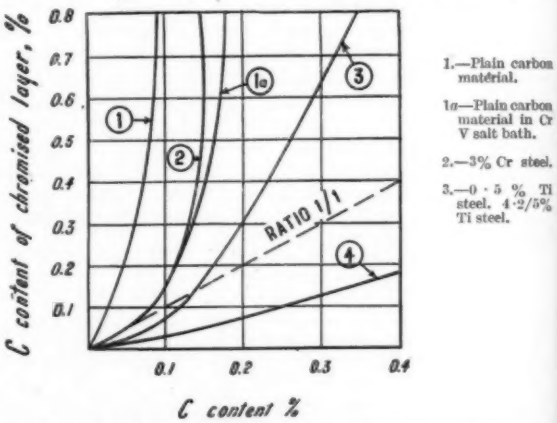


Fig. 3.—Relation between C content of material and chromised layer produced by gas-phase process.

The chief disadvantage of the gas-phase process is that its applicability is restricted to special materials. This is due to the aforementioned fact that the resultant diffusion of the carbon into the chromised zone impedes the diffusion of the chromium into the material, so that only thin chromised layers can be produced, which prove unsatisfactory. The relationship between the carbon content in the virgin metal and that in the chromised layer is charted in Fig. 3, which shows the characteristic curves obtained with the gas phase chromising of a 3% Mn steel, a 3% Cr steel and two titanium steels of 0·5 and 2·5% Ti content respectively.[3] The characteristic curves of a plain carbon steel obtained with both the gas-phase and a Cr-Va salt bath process are also shown. It will be noted that the inclusion of 2·5% Ti is so effective that the carbon content of the chromised layer has actually decreased below its original percentage.

3 H. Bennek, W. Koch, and W. Torfaute, *Stahl und Eisen*, Vol. 64, 1944, p. 265.

4 C. Wagner and V. Stein, *Zeitschr. f. Physik. Chemie*, vol. 192, 1943, p. 129.

5 C. Wagner, *Zeitschr. f. Physik. Chemie*, Vol. 192, 1943, p. 157.

Somewhat less effective, although still completely satisfactory with carbon contents up to 0·13%, is the action of 0·5% Ti. For carbon contents of up to 0·1% the 3% Cr and 3% Mn materials also give good results. The beneficial effects derived from the addition of vanadic chloride to a salt bath is exemplified by the respective curves given for the plain carbon material.

The addition of titanium to the steel is also of influence upon the depth of the chromised layer produced. If a certain Ti percentage is exceeded, the formation of δ—ferrite will serve to increase the speed of diffusion, and thus influence the depth of the layer. It is reported that with high Ti contents layers of up to 0·25 mm. depth can be produced by gas-phase chromising at 1,100° C. for 4 hrs. The following three materials are claimed to have proven particularly suitable for chromising in German practice :

0·1 C	0·5 Ti	Y.P.	21,300 lb./sq. in.	U.T.S. 48,280–59,640 lb./sq. in. Elongation 20%
0·1 C	1·5 Mn 0·5 Ti	Y.P.	56,800 lb./sq. in.	U.T.S. 78,100–99,400 lb./sq. in. Elongation 15%
0·1 C	3·5 Mn	Y.P.	63,900 lb./sq. in.	U.T.S. 92,300–113,600 lb./sq. in. Elongation 15%

of them favourably and others unfavourably. The employment of reaction-accelerators offers, however, the possibility of influencing both the reaction in the bath and the diffusion process in the metal. Referring to Fig. 4, it is seen that by the choice of different $CrCl_3/CrCl_2$ ratios in the bath the depth of the chromised layer can be greatly varied. Here it is shown that in the case of the wrought iron a greatest depth of chromised layer, amounting to some 0.14 mm., is produced with a $CrCl_3/CrCl_2$ ratio of 7, with a time treatment of 4 hrs. at 1,100° C. But in the case of the ferritic 5% silicon steel a $CrCl_3/CrCl_2$ ratio of 5 is required to produce a maximum depth of layer, which is shown to be some 0.26 mm. It may be mentioned that the greater layer depth in the silicon steel is due to the ferritic structure of this material.

When discussing the gas-phase chromising process, it was mentioned that the diffusion of the chromium into the material is hindered by the resultant carbon enrichment of the layer zone. It was further stated that this undesirable effect can also be counteracted by an appropriate composition of the bath, as, for instance, by the addition of vanadic chloride. This causes a small

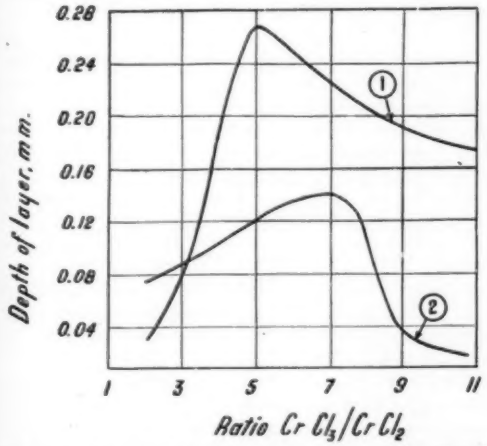


Fig. 4.—Influence of $CrCl_2/CrCl_3$ ratio in salt bath upon depth of chromised layer.
1.—5% Si steel. 2.—Soft iron.

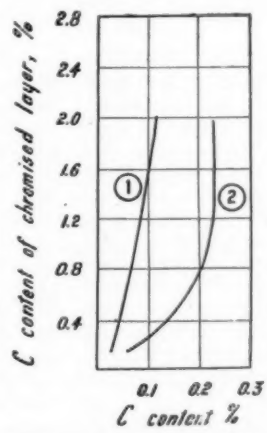


Fig. 5.—Influence of carbon content of material upon C content of chromised layer produced by salt bath method.
1.—Ordinary Cr salt bath. 2.—Improved Cr V salt bath.

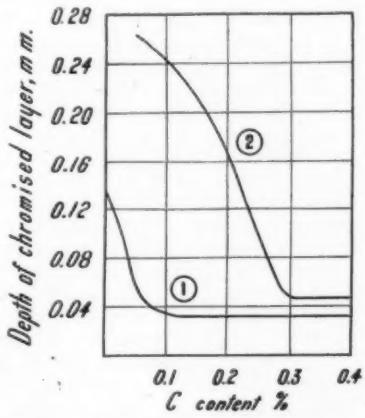


Fig. 6.—Influence of C content of material upon depth of chromised layer produced by salt bath method.
1.—Ordinary Cr salt bath. 2.—Improved Cr-V salt bath.

Salt Bath Process

By mixing molten chromous chloride with the chlorides of other metals, as, for instance, with borium, calcium or magnesium chlorides, the chromous chloride will remain stable in the bath up to temperatures of 1,200° C. While the concentration of the $CrCl_2$ in the vapour phase is governed by the factors of temperature and partial pressure, the salt bath method permits the maintenance of any desired concentration of the $CrCl_2$ up to some 50%. This is considered the main advantage of the salt bath method of chromising, as it enables the choice of the most suitable concentration for a given application. As a rule, a $CrCl_2$ concentration of some 20% will be found to produce sufficiently deep chromised layers with a conveniently short process time.

An important point to be considered is the fact that in technical steels the carbon, silicon, manganese, aluminium and titanium, etc., contents also react, some

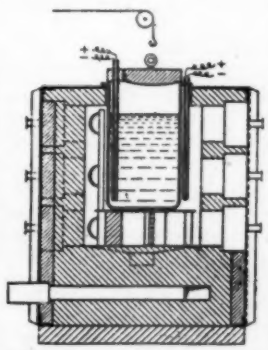


Fig. 7.—Latest type of gas-fired salt bath for chromising used in general practice.

amount of vanadium to diffuse into the metal together with the chromium, thereby impeding the diffusion of the carbon into the chromised zone. In this way the rise in the carbon content of the chromised layer can be considerably reduced, as illustrated in Fig. 5. Since the speed of diffusion of the chromium is not affected by the attendant diffusion of the vanadium, this also enables the production of a much deeper chromised layer. This beneficial effect of simultaneous vanadium diffusion is exemplified by

Fig. 6, which shows that in the case of a 0·1% C steel the chromised layer is increased six to seven-fold.

A gas-fired salt bath which is representative of latest German practice is outlined in Fig. 7. Where temperatures of 1,200° C. are to be employed, the bath container may be indirectly heated by immersing it in another electrically electrode-heated salt bath. It is reported[3] that with this temperature chromised layers of 1 mm. depth can be produced.

The Determination of the Hardness of Martensite and Austenite by means of the Microhardness Tester

By H. Hanemann

An investigation has been carried out to determine the hardness of martensite, austenite and their transition phases in a 1·7% carbon steel. The results of the investigation are given, together with a description of a new microhardness tester used for the experiments. The work is translated in the present article for which we are indebted to The Research Department, The Diamond Trading Co. Ltd., London, E.C.1.*

A NEW microhardness tester with an immersion lens is used for the determination of the hardness of martensite, austenite and of their transition phases in a carbon steel containing 1·7% C.

The Determination of Hardness Under the Microscope

The hardness of surface elements of microscopic dimensions can be determined by means of the microhardness tester[1]. This instrument uses a small four-sided diamond pyramid which is pressed into the material to be investigated with a force of 1–100 grams. The pyramid base is only 0·4 × 0·4 mm. in size, and the angle between the sides of the pyramid and its base, 22°—*i.e.*, the pyramid is similar in shape to the Vickers pyramid.

Experiments have been carried out in which the relationship between the load and the contact area of the indentation were determined with this instrument, (loads 1–100 gram, " microhardness determination "), and with the normal Vickers tester (loads of several hundred grams " microhardness " determination). They showed that the load/area relationship follows the same law for both instruments.

This is Meyer's Law[2] which states that :

$$P = a \cdot d^n \qquad (1)$$

where P denotes the load, a is a constant depending on the material to be tested, d the diagonal of the indentation (projection), and n another constant, also depending on the material to be tested.

Hardness is defined in both the Brinell and the Vickers tests as the ratio of the load to the surface area of the indentation produced. This is also used in the calculation of micro-hardness ; thus :

$$H = \frac{P \cdot 2 \cdot \cos 22^\circ}{d^2} \qquad (2)$$

If a material with n equal to 2 is tested, the same hardness number is obtained independent of the test load, and the microhardness is then equal to the macrohardness. Such a material follows Kick's law of similarity which states that for corresponding deformations in similar bodies of the same material the energy required is proportional to the volume of these bodies ; hence the specific pressure on corresponding areas are independent of the dimensions of these bodies. However, n is not always equal to 2 and usually is less than 2. For each load with such materials a different hardness number is obtained. Bernhardt[4] has developed some interesting and far-reaching theoretical hypotheses to account for such deviations from Kick's law.

Since there are materials, the hardness of which varies with the test load, it is necessary to agree either on the load or on the size of indentation with which the measurement is to be carried out. If we fix a load we would get in very hard bodies indentations which are too small, and in very soft bodies indentations which are too large. It is better, therefore, to standardise a certain size of indentation, and it is proposed to use for the microhardness determination indentations with (projected) diagonal sizes of $d = 5$, 10 and 20 microns. The corresponding hardness numbers will be denoted by $H_{5\mu}$, $H_{10\mu}$ and $H_{20\mu}$. Details of this procedure are given in a paper by Schulz und Hanemann[2].

In the measurements of martensite and austenite $H_{5\mu}$, and the value of n were determined. The value of a can then be calculated from the following formulæ :

$$H = \frac{1854 \cdot P}{d^2} \text{ kg/mm}^2 \text{ (P in gram } d \text{ in micron)} \qquad (3)$$

$$H_{5\mu} = 74 \cdot 16 \cdot P \text{ kg/mm}^2 \qquad (4)$$

$$a = \frac{H_{5\mu}}{74 \cdot 16 \cdot 5^n} \qquad (5)$$

Description of a New Microhardness Tester

For these measurements a newly developed microhardness tester was used, which differs from the one described earlier[1] in the design of the optical system.

* Communication from the Metallographical Institute of the Technische Hochschule, Berlin. Report No. 578 of the " Material " Committee, of the Verein Deutscher Eisenhüttenleute. *Archiv. f. Eisenhüttenwesen*, Vol. 15, 1942, pp. 403–406.

1 Hanemann, H., and E. O. Bernhardt, *Zeitschr. f. Metallkunde*, Vol. 32, 1940, pp. 35–38.

2 Schulz, F., and H. Hanemann, *Zeitschr. f. Metallkunde*, Vol. 33, 1941, pp. 124–134.

3 Meyer, E., *VDI, Zeitschr*, Vol. 52, 1908, pp. 645–654, 740–748, 835–844 ; *Forsch.-Arb. Ing.-Wes.*, Berlin, No. 65, 1909, pp. 1–61.

4 Bernhardt, E. O., *Zeitschr. f. Metallkunde*, Vol. 33, 1941, pp. 135–144.

2·5 2·5 5

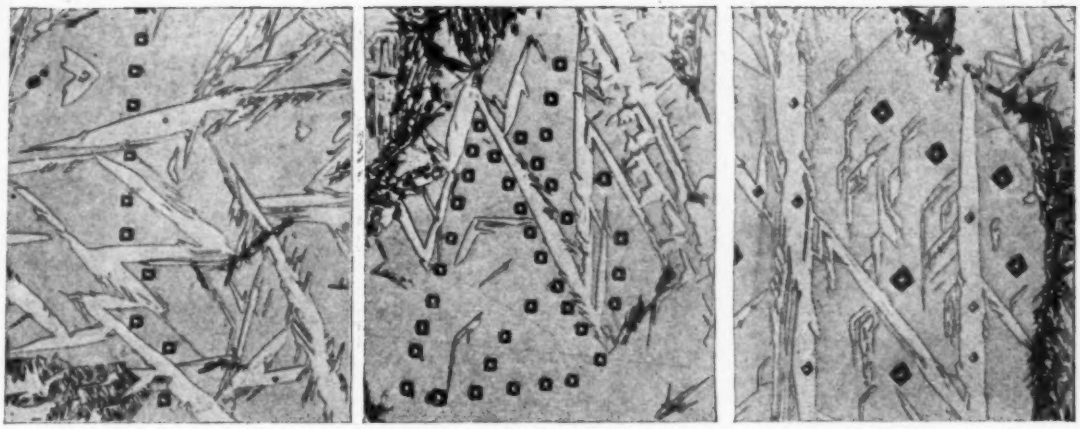


Fig. 2. Fig. 3. Fig. 4.

Figs. 2–4.—Microhardness indentations in quenched and untreated steel with 1·7% C, 0·14% Si and 0·45% Mn (Magnification ×600 ; etched with a 1% alcoholic solution of nitric acid. At the top of the illustrations : test load in grams.

The power of resolution of the optical system has been increased to the limit, the measurement of hardness can be carried out on areas as small as the wave-length of light. Objectives of such high power of resolution have to be used with immersion liquids and from the sine law follows that the back surface of their front lens comprises more than a hemisphere. The mounting of such a lens is difficult in ordinary microscopes, and it is usually carried out by cementing the lens along its edge to the lens mount. Such a lens is, of course, easily loosened by even slight pressure on its front. In the micro-hardness tester in which the front lens carries the diamond pyramid and is required to transmit the full test load this problem of mounting was solved in an unusual way. Fig. 1 shows a cross section through this front lens which carries the pyramid of approximately 0·4 mm. diameter in an axial hole. As this pyramid forms a stop for the axial rays the centre of the front lens is optically ineffective and therefore it is possible to utilise the central part of the back surface for the transmission of the test load. The lens system was so designed that the vertex distance between the first and second lens became negative. The second lens is then ground with a spherical hollow in which the first lens is supported. The mounting of the second lens is strong enough to transmit the test load of 0·5 kg.

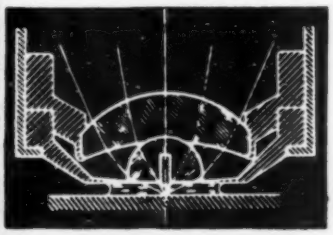


Fig. 1.—Mounting of the front lens of a new microhardness tester, the third and subsequent lenses of the objective have been omitted.

The remaining details of the new micro-hardness tester[5] are identical with those of the hardness tester previously[1] described.

The objective was designed as an Apochromat with a numerical aperture of 1·35 and a magnification of × 56 The aperture of the central stop formed by the diamond pyramid is 0·45. This central stop has little influence on the power of resolution as shown by Bernhardt[6]. From Abbée's theory the power of resolution is found to be 0·3–0·4 micron. These latter figures denote the least distance apart of the element of a regular structure of which separate images are formed by the optical system. They can be taken as a measure of the

TABLE I

MICROHARDNESS OF MARTENSITE, AUSTENITE AND THEIR TRANSITION PHASES IN A STEEL OF 1·7% CARBON CONTENT AFTER VARIOUS FORMS OF HEAT-TREATMENT

Sample No.	Time of tempering	Temp. at which tempered	Quenching medium	Martensite		Austenite	
				Const. n	Microhardness $H_{3\mu}$	Const. n	Microhardness $H_{3\mu}$
113 ..	Not heat-treated			2	868	2·34	239
81 ..	1	100		2	1,083	2·15	251
82 ..	2	100		2·04	1,157	2·27	230
83 ..	4	100		1·67	1,085	2·24	258
84 ..	8	100		1·61	1,135	2·21	246
85 ..	12	100		1·61	1,100	2·34	237
86 ..	16	100		1·88	1,172	2·28	231
87 ..	¼	150		1·89	1,113	2·14	313
88 ..	1	150		2·01	1,143	2·08	329
89 ..	¼	200		1·98	920	2·10	291
90 ..	1	200		1·91	801	2·18	280
91 ..	¼	250		2·14	772		
92 ..	1	250		2	750		
93 ..	2	250		2·04	727		
94 ..	4	250		1·97	757		
95 ..	8	250	Air	2·12	727		
96 ..	1	275		2·14	750		
97 ..	2	275		2·08	734		
98 ..	¼	325		2·03	683		
99 ..	1	325		2·08	697		
100 ..	2	325		1·88	675		
101 ..	4	325		1·86	697		
102 ..	8	325		1·91	668		
103 ..	¼	350		1·92	716		
104 ..	1	350		2·04	664		
105 ..	2	350		2·03	668		
106 ..	¼	400		1·89	631		
107 ..	1	400		2·02	620		
108 ..	¼	450		1·95	527		
109 ..	1	450		2·04	519		
114 ..	¼	450	Sand[1]	1·94	542		
110 ..	¼	500	Air	2·04	464		
115 ..	¼	500	Sand[1]	1·97	464		
111 ..	¼	600	Air	2·12	341		
116 ..	¼	600	Sand[1]	2·23	321		
112 ..	¼	650	Air	2·03	312		
117 ..	¼	650	Sand[1]	2·11	315		

[1]Cooled in hot sand.

5 Produced by C. Zeiss, *Jena*.

6 *Zeiss-Nachr.*, Vol. 3, 1940, pp. 280–286, 287–291.

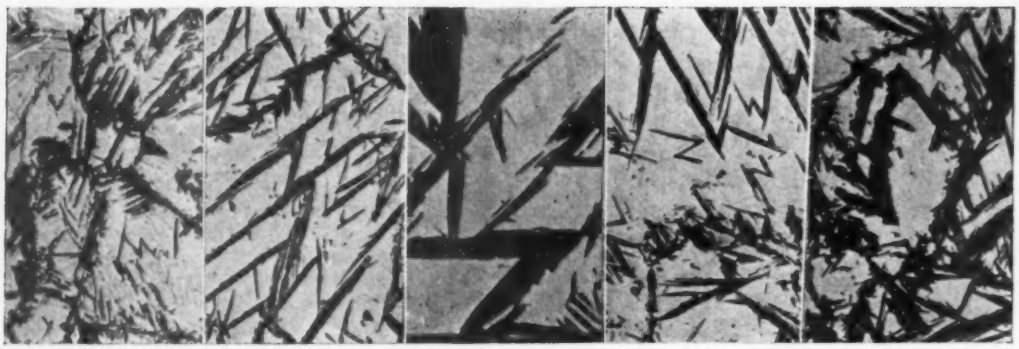


Fig. 5.—Tempered for 8 hours at 100° C. (× 200)

Fig. 6.—Tempered for 1 hour at 150° C. (× 200)

Fig. 7.—Tempered for ½ hour at 200° C. (× 1,000)

Fig. 8.—Tempered for 1 hour at 200° C. (× 200)

Fig. 9.—Tempered for 1 hour at 250° C. (× 200)

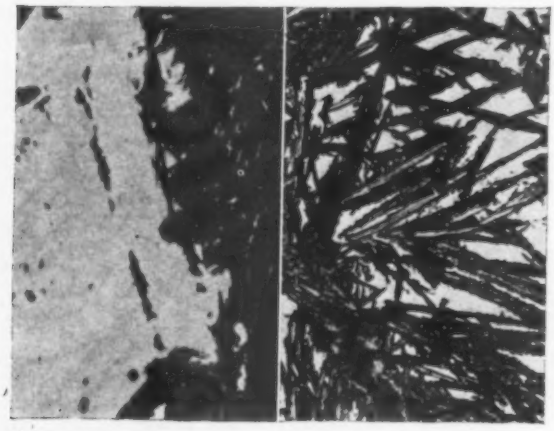


Fig. 10.—Tempered for 2 hours at 250° C. (× 1,000)

Fig. 11.—Tempered for 4 hours at 250° C. (× 200)

power of the objective to produce a geometrically similar image of a small object. The mean probable error in the measurement of a single hardness indentation is thus of the order of the wave-length of light, and it is therefore not possible with such small indentations to arrive at absolute values of hardness. Differences in the size of these indentations which are far below this limit of resolution, can still be measured, however. We can, therefore, with certainty diagnose differences in hardness, though we are not capable of giving absolute values.

To make full use about this power of resolution of the new objective it was necessary to grind the pyramid indenter so carefully that optical tests could no longer reveal a deviation from its prescribed form. The pyramid used for the tests was capable of producing indentation of definitely square shape, the sides of which measured less than 1 micron.[6]

Description of Tests

The samples used in these tests were slices of 8 mm. thickness cut from a 20 mm. diameter rod of a steel which contained 1·69% C, 0·14% Si and 0·45% Mn,

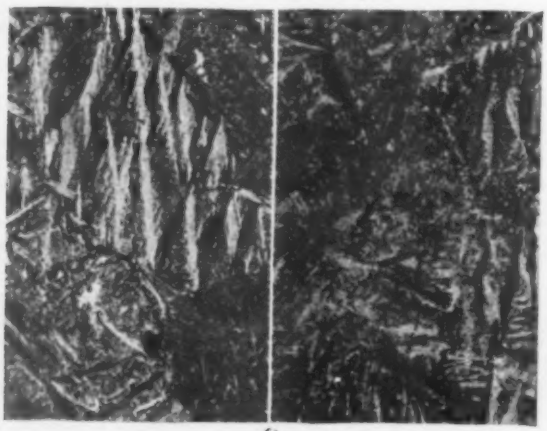


Fig. 12.—Tempered for 8 hours at 250° C. (× 200)

Fig. 13.—Tempered for 1 hour at 325° C. (× 200)

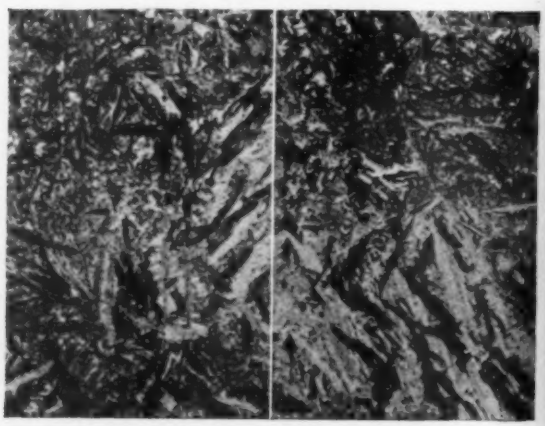


Fig. 14.—Tempered for 1 hour at 400° C. (× 200)

Fig. 15.—Tempered for ½ hour at 500° C. (× 200) (Sample cooled in hot sand.)

Figs. 5–9, 10–15.—Structure of the same steel after quenching and various forms of heat-treatment. (Etched with 1% alcoholic solution of nitric acid ; all samples except the last one were cooled in air after tempering.)

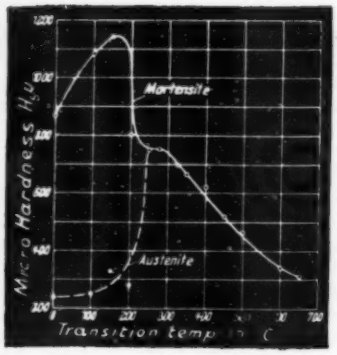


Fig. 16.—Micro hardness of martensite, austenite and of their transition phases in a steel of 1·7% carbon content, depending on the temperature reached in the tempering process. (Time of heat-treatment, 1 hour or ½ hour.)

were heated for 20 minutes at 1125° C. in an atmosphere of nitrogen and quenched in a 10% saline solution. The tempering temperatures and times are given in Table I. For tempering above 450° C., a nitrogen atmosphere was provided. Care was taken to polish the samples without heating them.

The immersion oil used in the hardness tests was of the usual kind but was always fresh, All indentations were measured immediately after being made. Martensite was tested with loads of 5, 10 and 15 gram, samples 103–117 being also tested with a load of 20 grams. In the case of austenite the loads were 2·5, 5, 10 and 20 grams. From 3 to 7 tests were made for each load according to the manner in which the measurements lacked consistency, and the mean taken as the value of the diameter of the indentation. In every case the time of loading was 30 sec.

The tests of martensite were carried out on small needles from which sufficiently uniform values could be obtained. If broad areas of martensite were tested, the values always varied considerably, presumably because these areas were thin flat crystals which were penetrated by the pyramid. Austenite and martensite were tested in several grains for each load. If the samples were tempered above 200° C., the austenite showed a fine distribution of fine, grey needles which were difficult to recognise. Above 250° C., these needles became so numerous that it was impossible to carry out a hardness measurement on the remaining austenite. (Compare Figs. 2–4 with Figs. 7–10.) The micro hardness of these fine, grey needles corresponded evidently to that of the dark grey needles of martensite. A systematic investigation of these needles was impossible as their width was too small for the instrument. The observations made would indicate that the final transition of the austenite is apparently preceded by a partial recrystallisation presumably by the formation of fine needles in martensite. If the temperature of 250° C. was exceeded the same hardness was found in places which were previously austenitic as those which were previously martensitic. Austenite thus changes above 250° C. into products of the same hardness as the transition products of martensite.

Micro Hardness of Martensite and Austenite

Figs. 2–4 show the structure of quenched samples and of samples not heat-treated. The hardness of martensite is considerably greater than that of austenite. (See Table 1, Fig. 4.) Within the austenite the hardness is uniform, both close to the martensite needles and further away from them. The carbon content of the austenite is thus apparently the same in both these areas. This would support the theory that the formation of martensite takes place without the diffusion of carbon through austenite.[7]

Figs. 5–15 show the structure of tempered samples. The hardness of martensite in quenched steel was 868 and this exceeds considerably that of austenite at 239 (Table 1). By heating for several hours at a temperature of 100°–150° C. the hardness of martensite increases to 1100, while that of the austenite changes only slightly. The transition of the tetragonal martensite to the cubic form is thus attended by a considerable increase in hardness. The simultaneous and slight increase in the hardness of the austenite is presumably due to continued precipitation of fine needles of martensite. Tempering at 250° C. causes a decrease of the hardness of martensite to 750, and an increase of the austenite transition phase to the same value. Such tempering thus makes the steel mechanically uniform, but with the retention of a high degree of hardness. Tempering at higher temperature causes a steady decrease of hardness. Fig. 16 shows how the micro-hardness of martensite and austenite varies with the temperature reached in the heat-treatment. The heating time was 1 hour for the lower temperatures and ½ hour for the higher temperatures.

The constant n was found exactly equal to 2 in the case of the tetragonal martensite; it is smaller for the cubic form. For the transition phases above 300° the value is again equal to 2 on the average. For austenite it exceeds 2.

7 Hanemann, H., U. Hofmann and H. J. Wiester, *Arch. Eisenhüttenwesen*, Vol. 6, 1932/33, pp. 199–133.

The Institute of Physics

AT the Annual General Meeting of the Institute of Physics, held on Monday, 4th June, 1945, the following were elected to take office on October 1st, 1945:—President: Sir Frank Smith; vice-president: Prof. A. M. Tyndall; honorary treasurer: Major C. E. S. Phillips; honorary secretary: Prof. J. A. Crowther; ordinary members of the board: Dr. B. P Dudding, Mr. A. J. Maddock, Prof. W. Sucksmith and Dr. C. Sykes. Prof. C. T. R. Wilson, F.R.S., was elected an Honorary Fellow of the Institute.

The *Report* for 1944 of the Institute shows that its membership increased by some 15 per cent.—a new record. Despite war-time difficulties the Groups and the Branches both at homes and overseas held meetings and conferences covering a wide range of technical and professional subjects. A new and very successful venture was the publication of a series of lectures given before the Industrial Radiology Group as a *Handbook of Industrial Radiology*. *The Report on the Teaching of Mathematics to Physicists*, prepared jointly with the Mathematical Association, was published during the year and was well received. Arrangements with the Ministry of Education for the establishment of National Certificates in Applied Physics were completed by the setting up of a Joint Committee to administer the scheme agreed. The *Report* also mentions many other matters which the Board of the Institute considered during the year, including Patent Law Reform, Higher Technical Education, the Limitation of Radio Interference and Professional matters

Aluminium Trains

THE range of applications for aluminium and its alloys was developing at a rapid rate immediately before the war and, although war conditions imposed restrictions on the use of the metal for other than special war purposes, partial release from these restrictions will permit a renewal of developments for peace time purposes. The fact that fabricating technique has greatly improved and that very many more competent operators are now available will assist in giving impetus to the further increase of applications.

Of the many fields open for development in the use of aluminium and its alloys, few offer greater scope in this country than railway transport. This was emphasised in the course of a discussion on a recent paper, before the Royal Society of Arts, by Mr. Christian Barman, F.R.I.B.A., "Public transport," said Mr. Barman, "is entering a period more eventful, more interesting and more exciting than any period since the building of our railways. English railways are the oldest in any country and most of our principal stations have been outgrown and are rapidly approaching obsolescence. Such an opportunity for re-designing our principal lines of public transport cannot recur for a long time."

He stressed the importance of the passenger in all transport considerations and, in connection with the railway station, claimed that heretofore it was the *train* that had mattered. Simplified travel, swift circulation of passengers, and the consideration of their comfort in every detail was only possible by good planning. Not only must travelling be made easier, but also more civilised.

A member of the audience suggested that the speaker had overlooked a most important item of construction in the form of light alloys. He instanced the streamlined trains in the U.S.A., built mainly from aluminium, citing one of the most modern trains on the Santa Fe route as an example of the proper use of light metals in connection with transport. He claimed that everything depends on the recognition in Great Britain of the importance of light alloys and electric power in connection with any plans for modern transport. Another speaker stressed the importance of the use of aluminium for interior decoration for land and sea transport and mentioned that aluminium and its alloys lend themselves to pleasant colouring. Keen interest was shown in these lighter metals by several members of the audience. Mr. Barman, in expressing agreement, pointed out that without exception every streamlined train in America uses light materials. Streamlining and light alloys are there almost synoynmous.

Machinery, Plant and Appliances (Control) Orders—

THE Machinery Licences Division of the Board of Trade has now been merged with the Industries and Manufactures (Engineering) Department. In future all applications for licences to supply controlled goods under the Machinery, Plant and Appliances Orders, and all related correspondence, should be addressed to the Assistant Secretary, Industries and Manufactures (Engineering) Department, Board of Trade, Millbank, London, S.W.1.

British Iron and Steel Federation

THE new constitution of the British Iron and Steel Federation which came into operation at the end of May, is based on membership by conferences representing manufacturers of the various products of the industry. Under the old constitution, which has now been superseded, the Federation membership was composed of individual firms and affiliated associations.

The largest of the 13 conferences, which will now constitute the Federation, is the British Steel Producers' Conference, comprising firms engaged in the heavy steel industry throughout the country.

The British Steel Producers' Conference have adopted a constitution and appointed as first chairman Mr. Ellis Hunter, deputy chairman and managing director of Dorman, Long & Co. Ltd., Mr. N. H. Rollason, managing director of John Summers & Sons Ltd., is deputy chairman of the Conference.

British Foundry School

AT a meeting of the Governing Body held recently in London (May 23) and attended by representatives of trade associations, research associations, scientific and professional institutions, technical and education committees, the Ministry of Education and firms connected with the industry—cast iron, steel, non-ferrous alloys, including aluminium, magnesium, bronze and brass founders—it was unanimously agreed to reopen the British Foundry School as soon as possible.

For the past two or three years there has been a steadily increasing demand that the British Foundry School should reopen. The Governing Body was unanimous in its decision to meet this demand at the earliest possible date. The necessity for the School in educating the foundry industry to a higher degree of efficiency, and in the creation of superior grade personnel, thus making the industry more attractive to labour, was stressed by one member.

Birmingham is regarded as the most suitable centre and the meeting felt the British Foundry School should be associated with an educational institution.

In its pre-war existence the School had had no failures; many ex-students are now occupying executive posts of considerable importance in the industry. It was confidently anticipated that ex-students would—in the light of their experience at the school—not hesitate to recommend as students those likely to profit by the course of instruction given.

A small committee of five members was appointed to consider ways and means of starting the School. The Committee will deal with such matters as finance, students, accommodation, staff, etc., and will report with all speed to the Governing Body, which will itself require modification following changes in the industry during the war.

HADFIELDS LTD., of Sheffield, announce that in connection with their re-organisation for post-war conditions, Mr. J. Hedley Jones has been appointed representative in South Wales.

Mr. Jones was Chief Engineer at their Hecla Works during the war years and is thoroughly conversant with Hadfield's steel products and mining and quarrying specialities. He will operate for the time being from :—1, Prospect Street, Newport, Mon.

Correspondence

The Shape of a Material's Reaction to Force

To the Editor, METALLURGIA.

Sir,

To tender, as I do, to Dr. Vivian my warmest congratulations on the cogency of his arguments in the articles which appeared in your March and April issues must inevitably imply a large measure of self-satisfaction with the close parallelism between our otherwise independent lines of thought, as evidenced by my own lesser contribution to your April issue. Nevertheless, I should like to be privileged to endorse most cordially Dr. Vivian's timely plea for a more fundamental approach to the very practical problem of the correct physical interpretation of engineering tests on materials which exhibit the stress-strain relationships characteristic of metallic crystalline aggregates.

In my own ponderings on this problem, I have found myself driven more and more to the thought, which is also implicit in some of Dr. Vivian's remarks, that the best approach to a satisfactory general solution is most likely to be found by striving to free ourselves from the hypnotic influence of the term " stress ; " and by recognising that, for the purpose involved, this term has no proper physical meaning apart from its connotation of the use of certain convenient but quite artificial external methods of measurement of the internal physical quantity that really concerns us, viz., the (stress-dimensioned) energy per unit of volume associated with some selected conditions of mechanical deformation. To some, this may sound like a distinction without a difference ; but I believe that the general literature very clearly indicates that to the majority of those who have practical occasion to think seriously about the present subject-matter, the suggested angle of approach will be novel enough to involve tuning-in on a hitherto unfamiliar wave-band. Most of us would have to confess, I think, that we have allowed ourselves to be stress-ridden for far too long.

As a case in point, brief reference may be made to the problem of fatigue. If, in Fig. 2 of my April contribution on " Indentation Hardness," we extrapolate the " strain-hardening curve " N'N below N, we shall clearly obtain a diagram of essentially the same form—although located by reference to a different range of numerical stress-values—as the familiar S–N (fatigue) curve for mild steel. This is apparently precisely what we ought to expect if we interpret fatigue curves as indicating the effect of initial differences in strain-energy content on the capacity of any given specimen to absorb additional strain-energy before fracture occurs. It would thus seem possible to contemplate a simple and direct correlation of fatigue (and alternating-bend) characteristics with primary stress-strain relationships (*see also* McGregor : *Journal of the Franklin Institute*, 1944, vol. 238, p. 170).

I am, Sir,

Yours faithfully,

Newport, Mon. G. C. RICHER.

May 25, 1945.

To the Editor, METALLURGIA.

Sir,

It is gratifying to have from Mr. G. C. Richer so emphatic an appreciation of the objects and arguments of the campaign I described in your March and April issues. If his " independent line of thought on parallel lines " involves, as it seems to, the substitution of strain-energy units of the mechanical properties for the present stress units, it has my sympathetic understanding, and I wish him luck in working things out on that basis. The pioneering steps in the campaign were probably the better for being on the usual stress basis. Two years ago, I vouchsafed the opinion that the foundations I was attempting to lay for the " renaissance " should be permanent enough until the whole was translated to an energy basis ; transference to the energy basis seems to be just what Mr. Richer wants to undertake.

I was certainly interested in Mr. Richer's application of a strain-energy theory to indentation hardness. It may be recalled that it was on the standard stress-difference theory that I founded my argument on indentation hardness for publication in the *Philosophical Magazine* of November, 1944. Yet, up to the last moment before offering my draft, I could hardly decide whether to employ a strain-energy argument I had developed, or the stress-difference argument. Not being quite satisfied with the mathematics of the former, I finally chose the latter for publication. If I could do some more work on the former, I should be glad to offer it to METALLURGIA. Incidentally, what might be described as a " maximum stress " theory was the basis of argument employed for greater simplicity and brevity in my explanation (which had to be much curtailed) in the Proceedings of the I. Mech. E., 1944, Vol. 151, No. 2, pp. 143–145.

It does not seem easy, however, to agree all Mr. Richer's points with those of my paper, " An Analysis of Hardness." In the first few lines of his article he refers to a widely recognised universal characteristic curve of mechanical deformation, thereby making me open my eyes in wonder as to whether my campaign had already proved successful or had, after all, been unnecessary ! Later on, however, he reverts to a certain degree of mistrust or recantation, as when he says, " at any rate up to the end of Stage B the stress-strain characteristics of any given material are substantially the same for both tension and compression."

In spite of a few difficulties, which have proved capable of solution, I have been faithful to my absolute belief that, with the exception of inaccurate stress-strain diagrams which we should not be discussing, stress-strain diagrams are from beginning to end true characteristics of mechanical deformation in simple tension, compression and shear. And I imagine such faithfulness has been rewarded, not only by the solution of the above-mentioned difficulties, but by the discovery in recent months that the true fracture stress of tension is identical with that of compression. Proof of this seems to have been forthcoming from my analysis of hardness ; for, having derived from this analysis by fair mathematical treatment a formula for calculating *stresses of failure* at any point of the specimen below the Brinell ball, it then becomes apparent that the maximum possible failure stress under the ball is identical with the true tensile failure stress, i.e., the compressive stress in the

metal in actual contact with the ball, the B.H.N. in short, is associated with a failure stress there of the same value as the true tensile stress of fracture. It could hardly be a different story in the case of Vickers or Rockwell testing.

I think the true fracture stress is Bragg's "definite physical limit," and I have the above reason for believing that the "saturation values" of the tensile and compressive fracture stresses are identical.

How true indeed the quoted saying of Confucius is! How much more suitably stimulating to the human mind of to-day would be a saying—"It is better to welcome a little additional light than to become too accustomed to the gloom."

I am, Sir,

Yours faithfully,

A. C. VIVIAN.

Bedford.
June 2, 1945.

The Editor, METALLURGIA.

Sir,

It is very good of Dr. Vivian to comment so fully on my brief outline of a hypothetical "strain-energy" interpretation of indentation hardness values; and I am happy to find that we can march together in fundamental unity in the renewed prosecution of a strangely retarded "campaign," which was apparently first embarked upon by Ludwik in 1909 (see Swift: METALLURGIA, January, 1945).

I freely admit that the opening gambit of my April article may justify Dr. Vivian's gentle implication that I have been guilty of sacrificing an element of truth; but I deemed it best, if only to conserve precious space, to start by assuming that I was preaching to the converted even though my closing remarks very clearly reveal that I am fully aware that there may very regrettably still be many who have become "too accustomed to the gloom" of the traditional load-elongation diagram.

Dr. Vivian can rest assured that there is no element of "mistrust or recantation" in my use of the qualifying phrase "at any rate up to the end of Stage B . . ." My only object here was to "accentuate the positive" by making no initial assumptions whatever concerning that portion of the true stress-strain curve which does not appear to influence the hardness values recorded within the standardised time limits of routine indentation tests.

With regard to Dr. Vivian's mode of analysis of the hardness problem, I must confess that I have found it a little difficult to understand just why it should be considered necessary to introduce the possibly misleading concept of "failure," the more so since it appears to have been very conclusively demonstrated that the phenomena of strain-hardening can be *permanently* induced by "elastic" distortion of the lattice not associated with "failure" of any kind (Wood: *Proc. Phys. Soc.*, 1940, vol. 52, p. 114). Incidentally, a further corollary from my own speculations could apparently be that the "strain-hardening curve" NN' (see Fig. 2, April article) should meet the fracture stress horizontal at the point of "brittle rupture"; this concept would seem to fit in very well with the suggestion made by Dr. Wood (*loc. cit.*) that there is no reason to suppose that "elastic" distortion need stop short of that point, even if there is no prior plastic "failures."

I agree that there would appear to be no theoretical reason why identity should not exist between the true tensile fracture stress and the "saturation value" of the compressive stress; and that close agreement may be recorded in practice is certainly indicated by the data of Taylor and Quinney (*Proc. Roy. Soc.*, 1934, vol. 143, p. 325). My "does not necessarily follow" in the penultimate paragraph of the April article) was intended to apply to observed values under arbitrary test conditions rather than to theoretical values; and it is encouraging to learn that Dr. Vivian has been successful in obtaining additional experimental evidence in support of the theoretical view.

I am, Sir,

Yours faithfully,

G. C. RICHER.

Orb Iron Works,
Newport, Mon.
June 7, 1945.

Revocation of Control Orders

Foundry Bonding Materials

THE Minister of Supply has made the Control of Foundry Bonding Materials (No. 2) (Revocation) Order, 1945, which revokes the Control of Foundry Bonding Materials (No. 1) Order, 1941, by which the disposal and acquisition of bentonite, colbond and fullers' earth, except for certain uses exempted under Direction 1 of the Order, were made subject to licence.

Shaped Diamond Tools

THE Minister of Supply has made the Control of Shaped Diamond Tools (No. 2) (Revocation) Order, 1945, revoking the Control of Shaped Diamond Tools (No. 1) Order, 1942, which prohibited, except under licence, the production, disposal and acquisition of shaped diamond tools.

Fluorspar

THE Ministry of Supply has made the Control of Fluorspar (No. 2) (Revocation) Order, 1945, which revokes the Control of Fluorspar (No. 1) Order, 1942, by which the disposal and acquisition of fluorspar containing more than 60% calcium fluoride was made subject to licence. All restrictions on the sale and purchase of fluorspar are thereby removed.

Copies of each Order, which came into force on 4th June, 1945, may be obtained from H.M. Stationery Office, York House, Kingsway, W.C.2, or through any bookseller, price 1d., on or after that date.

Cadmium

THE Minister of Supply has made the Control of Non-Ferrous Metals (No. 18) (Cadmium) Order, 1945, which comes into force on 8th June, 1945. This Order varies the Control of Non-Ferrous Metals (No. 6) (Cadmium) Order, 1941, by exemption from the operation of the maximum price provisions cadmium (whether virgin or secondary) in the form of ingots, sticks or rods, which is delivered or intended to be delivered outside the United Kingdom or on board ship for export from the United Kingdom. Enquiries concerning this Order should be addressed to the Joint Controllers, Non-Ferrous Metals Control, Grand Hotel, Rugby.

Microporosity in Magnesium Alloy Castings

The principal problem in casting magnesium alloys is microporosity which causes leakage when the castings are used for a number of purposes. The investigation, which is summarised in this article, was carried out to determine the primary causes of microporosity and to develop remedies. The author's conclusions and recommendations have resulted from facts observed in the study.

MAGNESIUM alloy castings are widely used in aircraft construction and other applications where their high strength/weight ratio is particularly advantageous, and they are used in many cases in vital highly stressed components. It is imperative that such parts be free from defects, and the most difficult problem faced by the industry in this connection is the occurrence of minute porosity which has been designated by the distinctive term. In order to determine the primary causes of microporosity, study the contributory factors governing its incidence, and if possible, develop remedial measures, an extensive investigation has been carried out by W. A. Baker, of the British Non-Ferrous Metals Research Association, a report of which is given elsewhere.*

Sand castings of magnesium–aluminium and other magnesium-base alloys were examined, and in some cases were compared with similar aluminium-base alloy castings. The work was done on a laboratory scale and the test castings were of simple form in order to minimise moulding variables and to facilitate examination of the castings. In many cases the castings were designed to show microporosity to a much greater extent than would normally be encountered in good foundry practice. By so doing it was possible to demonstrate the effect of various factors governing the incidence of the defect, but the results so obtained must be interpreted with some caution in relation to magnesium casting practice.

Examination of Castings.—The two main methods used in practice for the detection of microporosity in magnesium castings are fracture tests and radiographic examination. Porosity is revealed in fractures through porous areas by yellowish-brown stains. The staining is commonly very slight in as-cast material, but becomes most pronounced after heat-treatment. Fracture tests were only occasionally used in this investigation, for the obvious reason that they precluded further examination of the casting. Castings were examined (i) by X-rays to determine the distribution of porosity, (ii) by density measurements to determine the amount and in some cases also for more quantitative data on the distribution, (iii) by micro-examination to determine the form, and (iv) by tensile tests to determine the effect of the porosity. In most cases the strength of porous castings is expressed as a percentage of the strength of DTD bars cast from the same melts. In a few cases the actual strength is quoted, and in general the range of strengths observed in unsound castings is also quoted, since this is particularly important in showing the maximum effect in the least sound portions of the castings.

As a result of this study of microporosity in magnesium alloy castings Mr. Baker summaries the facts observed which have led to his conclusions and practical recommendations, which are presented in full.

* *Jour. Inst. Metals*, 1945, **71**, 185–204.

Summary

(1) Alloys with little or no freezing range show "secondary pipe" when inadequately fed. In alloys which freeze over a temperature range the porosity appears as intercrystalline cavities, small in relation to the primary dendrite size. Similar features have been observed in other non-ferrous cast alloys.

The notable feature of porosity in magnesium alloy castings is that the small cavities occur in local patches, usually with a specific orientation, *e.g.*, in sand castings the cavities are concentrated on planes normal to the mould wall.

(2) The porosity appears in castings from melts which are free from gas which would be evolved during solidification.

(3) Melts treated with hydrogen just before pouring give castings which show the same characteristic porosity; the amount of the porosity in a given casting may be increased, or alternatively the porosity may appear where none had been present when "gas-free" metal was used.

(4) Treatment of melts with sulphur dioxide, carbon dioxide, and carbon monoxide does not cause gas unsoundness.

(5) Gas may be absorbed in the mould as a result of reaction between the metal and moisture in the mould. It has been established that hydrogen enters some other non-ferrous metals in this way. The use of inhibitors, such as sulphur, etc., prevents the gas absorption provided the mould is not locally overheated by the passage of a large volume of metal.

(6) The primary dendrites in magnesium-base alloys are generally very much smaller than those in other non-ferrous casting alloys. The grain-size can be greatly increased by small additions of beryllium, to a lesser extent by additions of titanium, and also by holding the melt for prolonged periods near to the liquidus temperature. The porosity in such coarse-grained material is distributed differently from that in fine-grained material. Fine-grained sand castings contain very shallow patches of porosity normal to the mould wall and frequently in an annular zone or even extending to the mould wall; whereas in coarse-grained sand castings the porosity is concentrated at the centre of the section, the patches being much deeper and in many cases almost equi-axial.

(7) External restraint tends to promote the appearance of the characteristic patches of porosity, but the effect is very small indeed and is of secondary importance in comparison with that of other factors governing the incidence of microporosity.

(8) There is a deficiency of low-melting-point constituent in microporous areas. This is apparent under the microscope and is confirmed by chemical analysis. In certain castings concentrations of low-melting-point constituent have been observed by both X-rays and micro-examination, the concentrations having a similar form to the patches of porosity.

(9) Mechanical tests on a very large number of unsound castings have shown that local concentrations of porosity may cause very serious reductions in strength, but that the patches of porosity normal to the mould wall have a less serious effect on strength than might be expected from their form, provided that they are uniformly dispersed through the casting. Thus the strength of an unsound casting is determined more by the distribution than by the amount of porosity present.

(10) The distribution of porosity in an imperfectly fed casting is determined by the casting technique, and while this will vary with each individual casting certain factors are common to any particular case, and these have been the subject of study in the present work. It has been found that methods of pouring which ensure that a favourable temperature gradient is established during pouring have a very beneficial effect both in reducing the amount of porosity in the casting and also, more important, in producing a fairly uniform distribution of the patches of porosity present. Measurements of temperature gradients in castings have shown that owing to the relatively low thermal conductivity and capacity of magnesium alloys, freezing tends to proceed too fast from the mould face inwards and there is insufficient time for heat flow from the feeder to set up and maintain the temperature gradient necessary for feeding of the casting at the expense of the feeder. Measures which delayed solidification therefore improved the soundness of the casting in many cases.

Conclusions and Practical Recommendations

(A) The evidence indicates that freezing shrinkage is the primary cause of porosity, but that dissolved gas (hydrogen) will aggravate the defect. To avoid porosity due to gas two precautions are necessary : (1) moisture or corrosion products on the metal before melting must be avoided ; and (2) overheating of parts of moulds, *e.g.*, by the use of multi-gates, must also be avoided.

(B) The facts are consistent with the following picture of the solidification process. In the initial stages freezing shrinkage is compensated by flow of the small primary crystals and the remaining liquid towards the colder parts of the casting. This flow is termed " mass-feeding." A stage is reached, however, when the primary crystals grow and touch one another and so hinder the " mass-feeding " movement ; it is thought that at this stage masses of more or less coherent dendrites separate from other similar masses in their attempt to move towards the cooler part of the casting to compensate for freezing shrinkage. The discontinuities formed between the masses of dendrites therefore appear on planes at right angles to the direction of movement, which will generally be a direction parallel to the mould wall. These discontinuities may be, and probably are in many cases, healed initially by flow of liquid metal through the mass of primary crystals and residual liquid ; but eventually, when only a small amount of liquid remains, for example, 10 per cent. of the mass of the casting, the relatively small pools of residual liquid in the greater part of the casting draw feeding liquid from the larger pools formed when the discontinuities were healed. The result is that the cavities are concentrated on the planes where the primary dendrites separated comparatively early in solidification. If feeding is very inadequate it is possible that discontinuities may be formed by the separation of the primaries and that these may never be healed by flow of liquid metal ; but eventually the liquid surrounding these discontinuities will tend to flow farther away to feed the final freezing shrinkage, so that whether the breaks in the primary structure are healed or not they become the centres of local concentrations of porosity.

(C) The remedy for microporosity in magnesium castings must be sought by using a foundry technique carefully devised to ensure adequate feeding and rigorously adhering to the satisfactory technique developed for the job.

In the present work no very striking differences have been observed between the various alloys examined, but it is stated by foundrymen that in practice certain alloys give less trouble with porosity than others. It is appropriate to mention here that great care must be exercised in judging the relative merits of alloys of appreciably different compositions by X-ray examination. It was noted earlier, for example, that microporosity is more clearly revealed by X-rays when the alloy contains heavy metals, and the presence of such metals may therefore give the misleading impression that the casting contains more porosity than another casting of different composition.

To minimise the amount of microporosity in any alloy, the pouring method, pouring speed, and pouring temperature should be such that steep temperature gradients are set up in the required direction during pouring. These temperature gradients must be maintained during solidification by the provision of adequate feeders, and in this connection two points need to be stressed : (i) the efficiency of the feeder is increased if the method of pouring is such that the feeder cavity is preheated by the metal flowing into the mould proper ; and (ii) solidification in magnesium alloys cast in green sand moulds tends to be unduly fast from the mould face inwards, and measures which delay solidification and so increase the time available for heat flow from the feeder will in many cases improve the soundness of the casting. This is likely to apply to castings of thin section or to castings where the distance from the feeder to the remote part of the castings is large in relation to the cross-sectional thickness. The most convenient practical method of controlling the overall rate of freezing appears to be by careful control of pouring temperatures, intelligent use of chills, heating the mould cavity immediately before pouring, insulating the feeder heads, etc. It should be noted that unduly slow solidification is harmful for three reasons : (*a*) conduction of heat may proceed too far and smooth out the desired temperature gradient, (*b*) the reaction between the metal and moisture in the mould may be promoted, giving rise to gas unsoundness, and (*c*) the grain-size may be increased and the casting consequently weakened. On all counts, therefore, the rate of solidification should be as high as possible consistent with obtaining the desired temperature gradients in the casting. Proper chilling of parts farthest from the feeders will clearly assist to this end.

Metallographic Identification of Ferro-Magnetic Phases

By Miss E. A. M. Harvey, M.A., B.Sc.

Physics Section of the Brown-Firth Research Laboratories.

The "magnetic etchings" method for the metallographic identification of ferro-magnetic phases is perhaps not so widely known as it deserves to be. This brief description of the method and apparatus used is given so that those workers not familiar with the detailed procedure, but who are interested, for example, in the identification of ferrite in austenitic steels, may have available working details of the simple apparatus required.

THE positive identification of magnetic constituents in a non-magnetic matrix, or vice-versa, is not always easy to accomplish metallographically. This problem has been solved in many cases by the use of the colloid pattern technique. If the polished surface of a metallographic specimen be covered with a thin colloidal suspension of magnetic particles, the application of a magnetic field will ensure a visible concentration of the colloid over the magnetic areas, and a characteristic mosaic pattern may often be seen.

Experimental Data

The apparatus used was based on that described by Avery, Homerberg and Earnshaw Cook[1] and the diagram of the electro-magnet is reproduced in Fig. 1. The method of preparation of the colloidal magnetite (Elmore's method[2]), recommended by the above authors, was found to be unsatisfactory and was, therefore, modified slightly. Elmore's method is :—

"To make the new colloid, first prepare a coarse, flaky precipitate by Lefort's method; dissolve 2 grams of $FeCl_2\ 4H_2O$ and 5·4 grams of $FeCl_3\ 6H_2O$ (or equivalent amounts of the sulphates) in 300 cc. of hot water and add with constant stirring 5 grams of NaOH dissolved in 50 cc. of water. Filter to remove salt and excess sodium hydroxide. Rinse precipitate several times with water and finally once with 0.01 N HCl. Then transfer the precipitate to one litre of 0·50% soap solution and boil for a short time. The former precipitate will now have become entirely colloidal with the exception of a very small quantity of undispersed oxide which should be removed by filtering hot. It is interesting that the success of the method depends entirely on the peptisation of the precipitate with HCl before it is added to water containing soap which serves as a protective colloid. A drop or two of the colloidal magnetite placed on the magnetised specimen will give an ample supply of sol particles for forming a pattern."

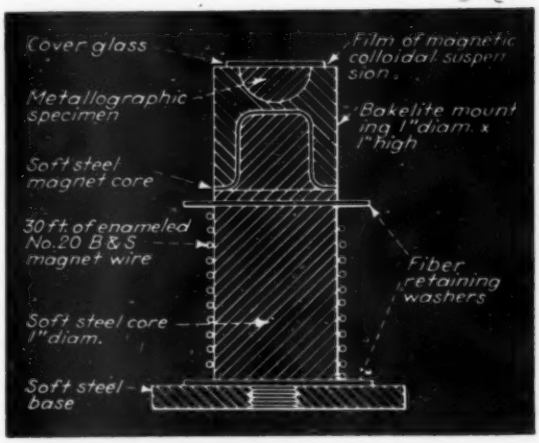
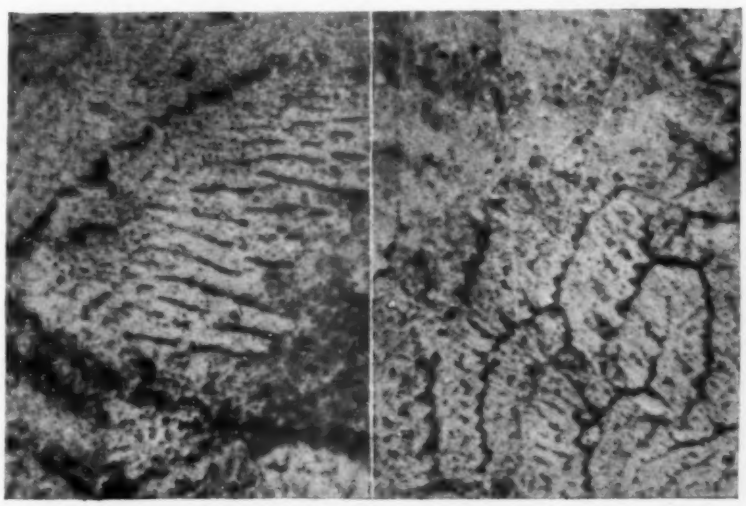

Cover glass
Film of magnetic colloidal suspension
Metallographic specimen
Bakelite mounting 1" diam. x 1" high
Soft steel magnet core
30 ft. of enameled No.20 B&S magnet wire
Fiber retaining washers
Soft steel core 1" diam.
Soft steel base

Fig. 1.—Electromagnet for producing colloidal magnetic patterns in metallographic specimens.

The colloids prepared in the above way did not appear to contain enough sol particles to give a good pattern. Various modifications were therefore

Figs. 2 and 3.—Colloid patterns on silicon iron. × 638.



1. Avery, Homerberg and Earnshaw Cook. *Metals and Alloys*, Nov., 1939, p. 353.

2. W. C. Elmore. *Physical Review*, 1938, Vol. 54, p. 309.

applied to the method of preparation and the best results were found to be obtained simply by boiling the final colloid solution until the volume was reduced to approximately three-fifths of that of the original volume.

Avery, Homerberg and Earnshaw Cook state that with 2,300 ampere turns a satisfactory field is produced although, in many cases, a weaker field produced better patterns, a method of control being desirable. In the present series of experiments satisfactory results were obtained with a current of about 3½ amps, which corresponds to approximately 400 ampere turns.

Materials Examined

A number of different materials have been examined by this method but the most interesting were a sample of silicon iron and specimens of 18% chromium, 8% nickel, 3% molybdenum, 1% titanium steel in various heat-treated conditions.

Fig. 4.—Colloid pattern on 18/8/3/1 steel (unetched) ×638.

Fig. 5.—Colloid pattern on 18/8/3/1 steel (etched) ×638.

The sample of silicon iron showed patterns similar to those published by Bitter[3] for large single crystals of silicon iron. Photomicrographs of the colloid patterns on two areas of the polished specimen are given in Figs. 2 and 3.

The colloidal patterns on the 18/8/3/1 steel samples showed a qualitative correlation with previous rough magnetic tests which had been carried out on the series in that no patterns were visible on any of the non-magnetic samples, while all the magnetic samples gave patterns. Figs. 4 and 5 show, firstly, the colloid pattern on the unetched specimen and secondly, the same area after etching for 5 secs. in "Newdip." These two photomicrographs illustrate very clearly the nature and high degree of fidelity of the information which may be obtained in this way.

Conclusions

The experiments described above show that :—

1. The necessary apparatus is relatively simple to construct.
2. The positive identification of ferro-magnetic phases is easy to accomplish, there being little or no difficulty in differentiating between the magnetic and non-magnetic phases.

Acknowledgment

This work has been carried out in the Brown-Firth Research Laboratories under the direction of Dr. C. Sykes, F.R.S., by whose permission this communication is published.

3. Bitter. "Introduction to ferromagnetism," p. 62.

WE are informed by the British Thomson-Houston Co. Ltd. that Mr. H. N. Sporborg, at the end of June, retired from the position of chairman of the board after completing 43 years' service with the Company.

Mr. Sporborg is also retiring from the positions of vice-chairman of Associated Electrical Industries Ltd., chairman of the Cosmos Mfg. Co. Ltd., chairman of Lamp Caps Ltd., director of Metropolitan-Vickers Electrical Co. Ltd., and director of The Switchgear Testing Co. Ltd.

Large Synchronous Condensers and Switchgear for New Zealand

THE British Thomson-Houston Co. Ltd. have recently received two important orders from the Government of New Zealand Public Works Department. One of these is for five 30,000 kVA. 750 R.P.M. synchronous condensers for operating on a 3-phase, 50 cycle, 11,000 volt supply. The equipments will be installed in North Island at the Public Works Departments' Otahuhu and Bunnythorpe Substations and will be used to control the voltage regulation of 220 kV. incoming transmission lines. These substations are the interconnecting points between the existing 110 kV. system and a new 220 kV. system which will bring power from a group of Generating Stations in the Upper Waikato River. The other order is for thirteen 600 ampere 110 kV. 2,500 MVA. oil circuit breakers, each with six bushing type current transformers and solenoid operating mechanism.

The synchronous condensers are rated for continuous loading over a range of 30,000 kVA. at zero power factor leading to 15,000 kVA. at zero power factor lagging, and they are also designed to take the full 30,000 kVA. at zero p.f. leading up to 11,750 volts and 15,000 kVA. at zero p.f. lagging down to 10,500 volts. Each condenser set includes main and service exciters and a slipring induction type starting motor operating from the 11,000 volt supply in series with the synchronous condenser windings. The machines are of the indoor type each complete with air cooling and water cooling equipments : a fire extinguishing equipment is to be supplied for each substation.

This order also includes the 11,000 volt solenoid operated switchgear and neutral earthing cubicles, control panels, relay panels, temperature indicating equipment and voltage regulating equipments for use with the condensers. The switchgear and control gear is arranged so that the whole sequence of starting operations is carried out entirely under automatic control following the operation of a "start" control switch.

War-Time Consumption of Non-Ferrous Metals

The Ministry of Supply has released detailed figures of the consumption during the war years of the metals within the scope of the Non-Ferrous Metals Control; they deal with copper, zinc, lead, tin, nickel, cadmium, antimony, cobalt and manganese. A full break-down into the main trades in which each metal was used is given for 1942, 1943, 1944 *and the first quarter of* 1945.

DURING the 5¼ years to March, 1945, the consumption (in virgin metal alone) of copper, zinc lead, tin and nickel amounted to a grand total of 4,861,000 tons. The division of this total was :—

	1940 to March, 1945 (inclusive) tons
Copper	2,254,000
Zinc	1,227,000
Lead	132,000
Tin	100,000

Practically all these tonnages were imported (either as ore or metal). Almost the whole of the lead and nickel, the major part of the copper and of the zinc, and a substantial part of the tin came from the British Empire.

These figures exclude the metal content of scrap which was very widely used. It is not possible to give the 1940 and 1941 totals for such scrap, but in the 3¼ years covering 1942 to March, 1945 (except for nickel scrap where the period is shorter) the metal content of copper, zinc, lead, tin and nickel in scrap (covering both "process" scrap and residues arising from current manufacturing operations and old and reclaimed scrap) amounted to 1,770,000 tons, viz. :—

	1942—March, 1945 inclusive (tons of metal content)	
Copper	1,004,000	
Zinc	440,000	
Lead	272,000	
Tin	32,000	
Nickel	22,000	(1944—March, 1945 inclusive).

The metals—whether in virgin form or as scrap—were used for ammunition and other war purposes, and for limited essential civilian purposes. Copper went mainly into cables and electrical equipment generally and into ammunition brass ; zinc into ammunition brass and diecastings for ammunition ; lead into cable coverings, batteries and ammunition ; tin into gunmetal, bronzes, solders and tinplate for foodstuffs ; and nickel into alloy steel. The full range of uses is shown in the accompanying tables.

Copper and Copper Alloys

U.K. Output of Main Copper and Copper Alloy Products —In Long Tons.

The following table shows the output in the U.K. of the main Copper and Copper Alloy Products in 1942, 1943, 1944 and the first quarter of 1945. Figures for the second quarter of 1945 will be published as soon as they are compiled. Comparable statistics for the years 1940 and 1941 are not available, but approximate estimates of the total consumption of virgin copper can be given, as follows : 1940, 440,000 tons ; 1941, 445,000 tons (comparing with 490,772 tons in 1942, 448,631 tons in 1943 and 348,139 tons in 1944, as shown in table I.)

Zinc

U.K. Consumption by Main Trades—in Long Tons.

The following table shows the consumption of Zinc in the U.K. by main trades in 1942, 1943, 1944 and the first quarter of 1945. Figures for the second quarter of 1945 will be published as soon as they are compiled. Comparable statistics for the years 1940 and 1941 are not available, but approximate estimates of the total consumption of new (virgin) Zinc can be given, as follows : 1940, 260,000 tons ; 1941, 256,000 tons (comparing with 258,564 tons in 1942, 222,785 tons in 1943 and 184,241 tons in 1944, as shown in Table II.)

TABLE I— COPPER AND COPPER ALLOYS

	Gross Output				Copper Content of Output			
	1942	1943	1944	1945 1st Qtr.	1942	1943	1944	1945 1st Qtr.
Unalloyed Copper Products—								
High Conductivity Rods and Strip	169,721	155,334	140,906	35,795	169,721	155,334	140,906	34,910
Rods and Sections	5,502	5,088	5,263	1,553	5,502	5,088	5,263	1,553
Strip and Sheet	34,986	30,549	28,484	7,951	34,986	30,549	28,484	7,951
Tubes	38,672	35,388	31,202	8,371	38,672	35,388	31,202	8,371
Castings	15,987	12,750	11,200	2,500	15,987	12,750	11,200	2,500
Sub-total	264,888	239,109	217,055	56,170	264,868	239,109	217,055	55,285
Brass and Other Copper Alloy Products—								
Wire	16,243	15,980	12,822	3,143	11,371	11,186	8,976	2,200
Rods and Sections	228,704	216,989	165,730	39,058	137,223	130,193	98,239	23,434
Strip and Sheet	419,092	410,586	264,304	44,968	293,892	287,412	185,013	31,478
Tubes	14,195	18,392	16,867	3,503	9,465	12,262	11,109	2,335
Castings	117,555	150,700	113,900	19,935	94,422	104,400	91,300	15,918
Sub-total	795,789	792,647	571,423	110,607	546,373	545,453	394,637	75,395
Copper Sulphate	38,850	43,260	45,387	13,468	8,862	10,815	11,347	3,368
TOTAL ALL PRODUCTS	1,099,507	1,075,016	833,865	180,245	820,103	795,377	623,039	134,048
for which:—								
Consumption of Virgin Copper	——	——	——	——	490,772	448,631	348,139	81,103
Consumption of Scrap Copper	——	——	——	——	329,331	346,746	274,900	52,945

TABLE II— ZINC

Main Trades.	U.K. Consumption of New and Scrap Zinc.			
	1942	1943	1944	1945 1st Qtr.
Brass	233,000	240,507	171,379	34,335
Galvanising: General	20,369	16,291	18,833	4,602
Sheet	4,742	955	3,570	2,315
Tube	14,611	10,885	9,400	2,254
Wire	1,584	1,425	2,800	1,600
Total Galvanising	41,306	29,556	34,603	10,771
Rolled Zinc	24,212	24,568	26,172	5,939
Zinc Oxide	25,376	26,961	33,136	8,224
Zinc Alloy Die-Casting	50,748	25,973	18,023	6,074
Zinc Dust	5,800	14,230	12,303	2,134
Miscellaneous Uses	15,632	14,058	13,300	2,883
Total All Trades:	396,074	375,847	308,916	70,360
of which:				
New Zinc: High Grade	159,554	122,897	96,535	23,727
Low Grade	99,010	99,888	87,706	21,684
Total New Zinc	258,564	222,785	184,241	45,411
Scrap (including Remelted) Zinc	137,510	153,062	124,675	24,949

Lead

U.K. Consumption by Main Trades—in Long Tons.

The following table shows the consumption of Lead in the U.K. by main trades in 1942, 1943, 1944 and the first quarter of 1945. Figures for the second quarter of 1945 will be published as soon as they are compiled. Comparable statistics for the years 1940 and 1941 are not available, but approximate estimates of the total consumption of imported Lead can be given, as follows: 1940, 250,000 tons; 1941, 215,000 tons (comparing with 229,462 tons in 1942, 197,053 tons in 1943 and 205,385 tons in 1944, as shown in Table III.)

TABLE III

Main Trades.	U.K. Consumption of Imported and Scrap Lead.			
	1942	1943	1944	1945 1st Qtr.
Cable	114,810	102,498	99,235	18,615
Battery (excluding Oxides)	27,133	26,989	29,210	7,296
Oxides and Compounds:				
Battery	20,970	23,617	25,118	6,334
Other Uses	19,384	19,102	23,303	6,020
White Lead	25,905	20,539	19,332	4,470
Sheet and Pipe	45,534	22,538	31,020	14,381
Ammunition	20,688	22,983	18,487	4,053
Foil and Collapsible Tubes	9,786	9,590	8,667	2,274
Solder	10,739	9,765	10,460	2,176
Alloys	10,146	9,466	7,235	1,811
Miscellaneous Uses	13,650	17,156	10,554	2,127
Total all Trades	318,745	284,243	282,621	69,557
of which:				
Imported Lead and Lead Bullion	229,462	197,053	205,385	51,517
Scrap Lead (including lead refined in the U.K. from scrap and home produced ores)	89,283	87,190	77,236	18,040

Antimony

U.K. Consumption by Main Trades—in Long Tons.

The following table shows the consumption of Antimony Metal and Compounds (in terms of antimony metal) in the U.K. by main trades in 1942, 1943, 1944 and the first quarter of 1945. Figures for the second quarter of 1945 will be published as soon as they are compiled.

Main Trades	U.K. Consumption of Antimony Metal and Compounds (in terms of antimony metal).			
	1942	1943	1944	1945 1st Qtr.
Antimonial Lead	1,921	2,184	1,833	534
Bearings	561	503	444	134
Fireproofing Camaflouge Scrim	Nil	68	701	10
Other Oxides and Compounds	1,307	1,146	1,300	344
Miscellaneous Uses	129	296	494	243
Total All Trades	3,918	4,197	4,772	1,265

The following figures exclude the consumption of antimony in scrap, which was as follows :—

	1943	1944	1945 1st Qtr.
For: Antimonial Lead	1,386	1,608	423
All other purposes	669	569	167
Total Consumption in Scrap	2,055	2,177	590

Nickel

U.K. Consumption by Main Trades—in Long Tons.

The following table shows the consumption of virgin Nickel in the U.K. by main trades in 1942, 1943, 1944 and the first quarter of 1945. Figures for the second quarter of 1945 will be published as soon as they are compiled. Comparable statistics for the years 1940 and 1941 are not available, but approximate estimates of the total consumption of virgin Nickel can be given, as follows: 1940, 23,500 tons; 1941, 23,000 tons (comparing with 21,751 tons in 1942, 17,218 tons in 1943 and 12,420 tons in 1944, as shown below).

Main Trades.	U.K. Consumption of Virgin Nickel.			
	1942	1943	1944	1945 1st Qtr.
Alloy Steels	16,842	11,947	8,186	1,262
Nickel Iron Alloys	720	1,071	893	131
Nickel Chromium Alloys	409	728	930	201
Malleable Nickel	610	741	618	250
Cupro-Nickel (Wrought Products)	1,872	1,452	685	167
Nickel-Silver (Wrought Products)	110	93	116	77
Non-Ferrous Foundries	76	77	60	18
Nickel Aluminium Alloys	556	526	380	36
Nickel Anodes	549	565	537	280
Miscellaneous Uses	7	17	15	9
Total All Trades	21,751	17,218	12,420	2,431

The following table relates only to virgin Nickel. The following figures give the consumption of *Nickel in scrap* :—

	1944	1945 1st Qtr.
For: Alloy Steels	15,378	2,778
Other Uses	3,273	831
Total All Trades	18,651	3,609

Tin

U.K. Consumption by Main Trades—in Long Tons.

The following table shows the consumption of Tin in the U.K. by main trades in 1942, 1943, 1944 and the first quarter of 1945. Figures for the second quarter of 1945 will be published as soon as they are compiled. Comparable statistics for the years 1940 and 1941 are not available, but approximate estimates of the total consumption of virgin Tin can be given, as follows: 1940, 34,500 tons; 1941, 34,500 tons (comparing with 23,478 tons in 1942, 17,631 tons in 1943 and 18,435 tons in 1944, as shown below).

Main Trades	U.K. Consumption of Virgin and Scrap Tins			
	1942	1943	1944	1945 1st Qtr.
Tinplate	5,572	2,416	3,912	1,098
Solder	7,937	6,982	7,168	1,544
Alloys (Whitemetal, Bronze, Gunmetal, etc.)	14,158	15,392	14,738	2,999
Foil and Collapsible Tubes	1,613	734	608	115
Tinned Copper Wire	1,044	716	688	125
Tin Compounds and Salts	342	201	216	46
Tinning	1,006	666	652	148
Miscellaneous Uses	655	616	611	150
Total All Trades	32,327	28,723	28,593	6,225
of which:				
Virgin Tin	23,478	17,631	18,435	3,949
Tin in Scrap	8,849	11,092	10,158	2,276

(*Continued on page* 79)

High-Tensile Steel for Castings

By W. West, C. C. Hodgson and H. O. Waring

A number of aspects of steel making and foundry practice, investigated by the authors during the last five years in connection with the production of medium and high-tensile steel castings, are co-ordinated in a recent paper presented to the Lancashire Branch of the Institute of British Foundrymen. The types of steels were of the low- and medium-alloy varieties capable of giving ultimate tensile strengths from 35-65 tons/sq. in. on sections up to 4 in., or as high as 75 tons/sq. in. on smaller sections, accompanied by good ductility and resistance to impact. The paper is reproduced by courtesy of the authors.

IN the early days of the war the authors were called upon to organise the production of steel castings to meet the pressing demands of the Services. The rapid solution of problems encountered in their production was of immediate importance. To facilitate the work a number of investigations have been carried out on various aspects of steelmaking and foundry practice during the last five years and, although time did not permit these investigations to be concluded in the way the authors would have liked, the results obtained indicate a high degree of success in the production of medium- and high-tensile strength castings. In this paper an endeavour is made to co-ordinate this series of investigations.

The types of steel dealt with are low- and medium-alloy varieties capable of giving ultimate tensile strengths from 35–65 t.s.i. on sections up to 4 in., or as high as 75 t.s.i. on smaller sections, accompanied by good ductility and resistance to impact. It is proposed to deal broadly with the steel making practice, and the influence of chemical composition on mechanical properties and heat-treatment, and to include some notes on sands and methods of feeding. The steels around which these notes are written include carbon manganese steel containing small amounts of nickel, chromium, and molybdenum (referred to in the paper as "residual alloys") which, together with the manganese present, resulted in a steel approximately equivalent to a straight carbon steel containing 1½% manganese; manganese-molybdenum (1½% Mn) and chromium-molybdenum (1% Cr) containing similar amounts of residuals; non-standard steels containing larger amounts of residuals; and 2¼% nickel-chromium-molybdenum steels.

Plant and Equipment

The authors were not particularly fortunate in the steel making furnaces available; these consisted of basic-lined electric arc furnaces with conducting hearths, one a Greaves Ethells and the other an Electro-Metals, these had many years' service to their credit, and had been standing idle for a period of about five years prior to the outbreak of war. The conditions for steel making were less favourable than in more modern furnaces, and particular care was needed to ensure the high quality of steel required. On the heat-treatment side conditions were more favourable as a number of modern furnaces of different types were available. Sand handling during the period covered by the paper was on the batch principle, but the foundry is now being mechanised for reclamation, milling, and conveying.

Steel Making Practice

The turnings and scrap used come from the engineering departments of the company with which the foundries are associated and carry a certain amount of alloying elements which remain in the steel after melting. The nickel content in the bath usually varies between 0·5 and 1·0 per cent., and the steel after slagging also carries small amounts of chromium and molybdenum. For many foundry purposes these residual alloys would be considered a distinct disadvantage but for the type of work discussed here they have proved to be helpful. Charges consist of three types of material, foundry scrap (headers, etc.), medium- and light-weight steel scrap, and turnings, used in approximately equal amounts.

Generally speaking there are two ways of operating basic electric arc furnaces working on the two slag process. One method is to melt the charge and boil out the carbon to the required figure and remove the oxidising slag before any deoxidising additions are made. The other is to start the "killing" reactions by adding some form of deoxidiser (generally ferro-manganese or silico-manganese) before the oxidising slag is removed. The authors have used both processes extensively and experimented with several variants of the second one, and, whilst they agree that excellent steel can be made by either process, it is their experience that steel of the highest quality is more likely to be obtained consistently when deoxidation is commenced early in the heat than when left until after the oxidising slag has been removed, or the second slag added. As far as possible all operating conditions have been standardised, commencing with the make up of the charge, through melting and refining to teeming. Whatever method is used the correct place to deoxidise basic electric steel for castings is in the furnace and no amount of subsequent dosing will fully make up for deficiencies there, conversely, if the job is done properly in the first place the steel does not need dosing afterwards. The addition of aluminium to the metal as it leaves the ladle is not permitted; the practice still has its uses but so far as basic electric steel for castings goes it is an anachronism.

Hydrogen and Flakes

The phenomenon variously referred to as hair-cracks, flakes, and snow flakes is sometimes met with in forgings made from alloy steels. According to the literature there may be many causes for this defect and workers in that field are by no means unanimous about the real cause, but the majority of informed opinion inclines to the view that it is to be attributed to the presence of hydrogen in the steel. The authors have occasionally found tensile test pieces representing alloy steel castings to exhibit a peculiar bright patch in the fracture which appears to be a defect in steel castings analogous to flakes in steel forgings. A typical fracture of the type referred to is shown in Fig. 1. The serious way in which the properties of the steel can be affected are well illustrated by the results in Table I.

In considering measures to be adopted for the elimination of this defect it was assumed that it was a manifestation in alloy steel castings analogous to the flakes in steel forgings. Working on the most favoured opinion that such defects were due to hydrogen it became necessary to consider the means by which excessive amounts of hydrogen might find their way into the steel and to take steps to prevent this, or to remove it. The slag making material used at this time was burnt lime; and it is almost impossible to prevent a certain amount of hydration taking place, such lime could have carried hydrogen into the steel. The defect occurred most often in steels in which the carbon was fairly high and in heats which had not received much boiling during the oxidation stage. Two factors were investigated, burnt lime was replaced by limestone, and the steel was boiled vigorously so as to bring the carbon in the bath to a figure not exceeding 0·15% before slagging. When these measures were adopted the defect disappeared. Limestone gives rise to the evolution of a large amount of carbon dioxide gas, and in making the second slag the rapid evolution of this large volume of gas caused a certain amount of discomfort to the furnace crews, it was decided, therefore, to return to the use of burnt lime but to continue boiling out to a carbon content around 0·1%. When this procedure was adopted there was no further appearance of the bright, shiny type of flake or fissure, from which it was concluded that the vigorous boiling during the oxidising stage had eliminated the cause of the defect, and that the use of burnt lime, even if somewhat hydrated, had no adverse effect in this direction when used for the making of the second slag.

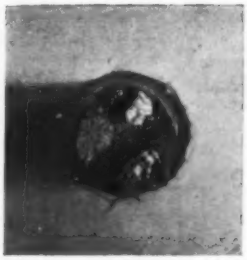


Fig. 1.—Showing bright patches in the fracture attributed to the presence of hydrogen in the steel.

TABLE I.

MECHANICAL PROPERTIES OBTAINED FROM TEST PIECES SHOWING FLAKED AND SOUND FRACTURES.

CAST A3028.
C 0·21, Si 0·45, Mn 1·58, S 0·012, P 0·020, Ni 0·75, Cr 0·13, Mo 0·28.
Specimens 1¼" × 1¼" × 4½".
Normalised 900°C. Oil Hardened 850°C. Tempered as shown.

Tempered	550°C.		600°C.	
Fracture	Flaked	Sound	Flaked	Sound
Y.P. (t.s.i.)	—	47·7	35·6	40·0
U.T.S. (t.s.i.)	45·4	54·4	44·2	48·0
E %	2	14	5	18
R/A %	10	26	20	40
Izod ft. lb.	—	18, 17	—	42, 40
U.T.S. E.	47·4	68·4	47·2	66·0

Aluminium Additions

(i.) *Control of Inherent Grain Size:*

The advantages of a controlled inherent grain size have long since been recognised in connection with wrought steel because of the vast amount of work which has been done following the initial investigation of McQuaid and Ehn on "abnormality"[1], and for most purposes a fine inherent grain size is preferred. The importance of inherent grain size in steel for castings has not received the same attention. The authors' experiments were made at a time when the castings weresupplied in the annealed or normalised and tempered conditions.

A knowledge of the McQuaid-Ehn grain size may be of value as an indication of the probable behaviour of the steel under a given set of conditions, but it is important to know the effect of variations in steel making practice on the structure of the steel in the condition in which it will be used. Usually, the smallest grain size and the most uniform ditribution of peralite and ferrite grains are desired. The grain size depends chiefly on the temperature of treatment, rate of cooling, and austenite-coarsening characteristics of the steel. The even distribution of pearlite and ferrite is directly related to the homogeneity of the steel, the segregation which takes place during the solidification of the steel usually leads to uneven distribution of the ferrite and pearlite after annealing, the slower the cooling after annealing the more pronounced the separation into pearlite-rich and ferrite-rich zones.

For some purposes a large addition of aluminium has been advocated as a primary deoxidiser, our experiments compared such an addition with ladle additions and with heats to which no aluminium was added. The conditions were, no aluminium addition, additions of one and of three lb. per ton added to the metal stream in the ladle, and the addition of 6 lb. per ton added under the oxidising slag immediately before slagging. The McQuaid-Ehn grain sizes, determined under standard conditions, are shown in Table II.

TABLE II.

EFFECT OF ALUMINIUM ADDITIONS ON MCQUAID-EHN GRAIN SIZE.

Cast Serial	Aluminium Additions	McQuaid-Ehn Grain Size
2976 ..	No aluminium addition.	4 and 5 predominating, a little 2.
2981 ..		4 and 5 predominating, a little 3.
3038 ..		4 and 5 predominating, a little 3.
2989 ..	1 lb. per ton added to metal stream in the ladle.	4
3034 ..		4 and 5 in about equal amounts.
3148 ..	3 lb. per ton added to metal stream in the ladle.	7 predominating, a little 8.
3150 ..		7 predominating, a little 6.
3170 ..		5 and 6 with a little 4.
3171 ..		6 with a little 5.
3001 ..	6 lb. per ton added to the metal under the oxidising slag as primary deoxidiser.	7 and 8.
3031 ..		7 with a small amount of 5 and 6.
3166 ..		7 with a little 6.
3168 ..		7 with a little 8.

When the structures of the annealed steels were examined under the microscope it was easy to draw a distinction between steel to which, on the one hand, none, or only a little, aluminium had been added, and, on the other, steel to which relatively large aluminium additions had been made; but it was not possible to distinguish between heats to which 3 lb./ton of aluminium had been added in the ladle or 6 lb./ton added in the furnace; sometimes the one and sometimes the other gave the finer structure. There was no doubt about the influence of 1 lb./ton and 3 lb./ton added to the ladle; the latter gave the smaller individual grains of pearlite and ferrite, although there was not much to

1 McQuaid-Ehn. The Effect of Quality of Steel on Case-Carburising Results. T.A.S.M. & M.E., 1922, vol. 67, p. 341.

choose between the two in the matter of general arrangement and size of the cellular formation of ferrite "kernels" and pearlite network in instances where this was the type of structure present.

Concurrently with this work the influence of these additions on mechanical properties was determined, and without going into the results of this work in detail the conclusions were that the use of 6 lb./ton of aluminium as a primary deoxidiser, without any subsequent additions, caused the steel to be slightly tougher than steel of similar analysis to which no aluminium had been added, or to which 1 lb./ton had been added in the ladle; but when comparing the addition of 6 lb./ton used as primary deoxidiser with 3 lb./ton added in the ladle an advantage in favour of ladle additions was indicated.

The general conclusion, therefore, was that a large ladle addition of aluminium to basic electric steel is an advantage in helping to produce a fine grain size and in increasing toughness, and, in general, produces results which are superior to those obtained when an even larger addition is used as a primary deoxidiser at an earlier stage in the steel making operations.

(ii.) *Control of Sulphide Distribution:*

The preceding experiments were made without reference to the previously published work on the influence of aluminium additions on the nature and distribution of sulphide inclusions and it is a remarkable coincidence that the addition decided on, 3 lb./ton, is around the optimum for obtaining high ductility.

Most steel founders must occasionally have had the misfortune to make a cast of steel which was later found to possess ductility less than that usually associated with its ultimate tensile strength. The condition does not persist after forging; steels of similar analysis which vary markedly in ductility in the cast and annealed and the cast and normalised conditions usually possess almost identical ductility after forging, as illustrated by the results shown in Table III.

No form of heat-treatment which the authors have found will remove this "dry" condition from steel in the cast condition if the heat of metal is predisposed to this defect, even such drastic treatments as homogenising at temperatures as high as 1100° C left the low elongation steel B1626 (shown in Table III) almost unaffected.

It is now fairly generally agreed that this type of defect is due to the difference in size and arrangement of the non-metallic inclusions present, particularly the sulphide inclusions. Typical inclusions on cast numbers B1626 and B1628 are shown in photomicrographs Fig. 2 and Fig. 3 each at a magnification of 500. The metal-inclusions interface area in the former must be many times that in the latter, for example, about 1,000 of the smallest inclusions shown in Fig. 2 could be packed into the largest of the inclusions shown in Fig. 3 and would of course have a surface area about 1,000 times as great.

TABLE III.

COMPARISON OF TWO STEELS GIVING "DRY" AND DUCTILE FRACTURES.

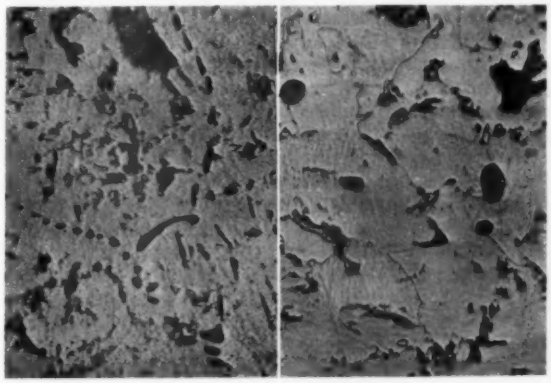
CAST B1626.
C 0·21, Si 0·22, Mn 0·90, S 0·033, P 0·027, Ni 01·00, Cr 0·12, Mo 0·09, Cu 0·09.

CAST B1628.
C 0·18, Si 0·19, Mn 0,90, S 0·029, P 0·020, Ni 0·98, Cr 0·13, Mo 0·11, Cu 0·19.

Normalised 900/925°C. Tempered 640°C.

Condition	Castings*		Forgings†	
Cast No.	B1626	B1628	B1626	B1628
Y.P. (t.s.i.)	24·0	23·0	27·8	26·8
U.T.S. (t.s.i.)	32·0	31·6	35·2	33·4
E %	19	31	32	33
R/A %	—	—	63	69
Izod ft. lb.	—	—	68/67	75/80

* Test Bars 1½" × 1½" × 4½" when heat treated.
† Test Bars ¾" dia. when heat treated.



Figs. 2 and 3.—Typical inclusions from two casts. ×500.

This connection between the distribution of the non-metallic inclusions and the ductility of the steel was referred to at some length by Sims and Lilliequist[2] as long ago as 1932 and since then a considerable amount of work has been done on the subject notably by Sims and Dahle [3, 4] and Crafts, Egan and Forgeng[5]. Views expressed on the problem, particularly by contributors to discussions, are by no means unanimous but it does seem that in the matter of adding aluminium to well killed steel for castings, either it should be given none or it should be given a really good dose. There appears to be slight evidence that an addition of calcium-silicon at the same time as a large addition of aluminium is beneficial in producing the right kind of non-metallic inclusion. The melting procedure adopted by the authors (originally for reasons which had nothing to do with the distribution of inclusions) is the same as that recommended by some of the American workers referred to in their work on inclusion distribution and ductility. We can confirm that with our present methods of slag making and finishing, test pieces possessing a "woody" or "dry" fracture are very rare occurrences and when found can generally be associated with something slightly abnormal in the slag condition. Whether this is to be interpreted as meaning that this has altered the amount of aluminium entering the steel we have not investigated.

In connection with this and allied problems it would be most helpful if there were available really satisfactory methods for the determination of small amounts of aluminium and alumina in steel. Some analysts may be inclined to protest that such methods do exist, but it is our experience that if this is so knowledge of them is not very widely disseminated.

2 C. E. Sims and F. B. Lilliequist. Inclusions, Their Effect, Solubility and Control in Cast Steel. A.I.M.M.E., *Technical Publication* No. 453, 1932.
3 C. E. Sims and F. B. Dahle. Effect of Aluminium on the Properties of Medium Carbon Cast Steel. T.A.F.A. 1938, vol. 46, p. 65.
4 C. E. Sims and F. B. Dahle. Comparative Quality of Converter Cast Steel. A.S.T.M., 1942, vol. 42, O. 532.
5 W. Crafts, J. J. Egan and W. D. Forgeng. Formation of Inclusions in Steel Castings. A.I.M.M.E., *Technical Publication* No. 1184. Metals Technology, April, 1940.

Temperature Measuring

Empirical methods for assessing the temperature of molten steel can yield good results when used by experienced observers, but it is recognised that more exact methods are desirable, and during the last few years much attention has been paid to methods suitable for measuring the temperature of molten steel, notably by The Liquid Steel Temperature Sub-Committee of the Iron and Steel Institute Heterogeneity Committee.[6, 7]

Optical pyrometers are useful instruments for determining temperature after the metal has left the furnace; they are not satisfactory for temperature measurements of the metal in the furnace, although it is obviously of more importance to know the temperature at this stage rather than at some later stage when nothing can be done to increase the temperature of the metal and it is possible to take steps which are capable only of reducing the temperature.

It now seems fairly well established that the most satisfactory thermocouple for use in steel melting furnaces is the platinum, platinum-rhodium combination. A considerable step forward has been achieved by Tinsley Industrial Instrument Co. who have developed an apparatus which amplifies the small e.m.f. produced by the couple and enables the temperature to be recorded by means of a pen and ink mechanism on a quickly moving chart.[8]

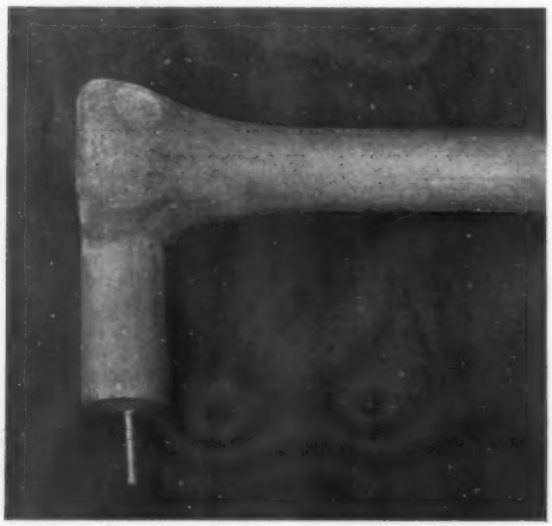


Fig. 5.—The immersion thermocouple ready for use.

The thermocouple e.m.f. operates a mirror galvanometer and the beam of reflected light falls on a photocell. The degree of illumination of the photocell controls the rectified current output of a thyratron valve, and by an ingenious manipulation of this current in a resistance in series with the galvanometer, it is rendered proportional to the input voltage (i.e. the e.m.f. of the thermocouple). The anode current through the thyratron valve is supplied from a 250-volt. section of a mains transformer, and it is used to operate a robust moving coil type galvanometer with the pen and ink mechanism recording in the moving chart. The amplification is independent of mains fluctuations, characteristics of valves, and other components. The apparatus is reliable and requires only slight periodic adjustments which can be carried out quickly and checked. The temperature of molten steel can be measured in five to ten seconds.

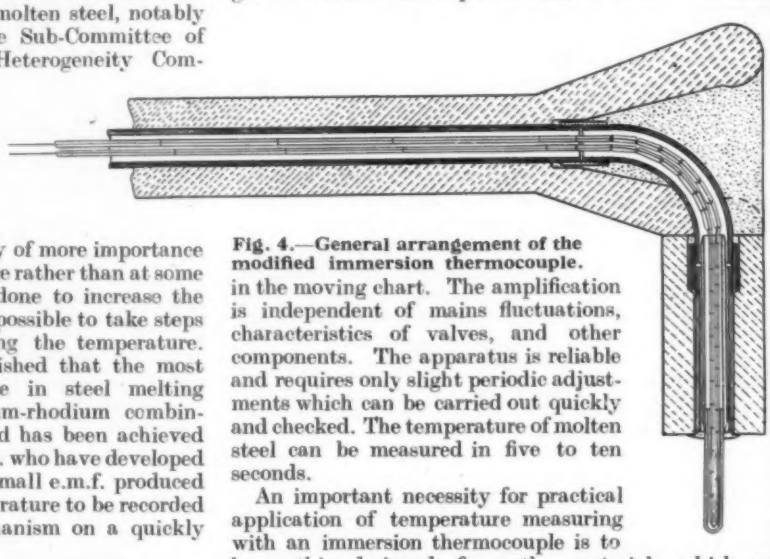


Fig. 4.—General arrangement of the modified immersion thermocouple.

An important necessity for practical application of temperature measuring with an immersion thermocouple is to have this designed from the materials which are readily available and constructed in such a way that repairs and maintenance can be quickly and cheaply done. We have developed such a thermocouple covering which can be made from materials which are to be found in most foundries or are readily obtainable. The general arrangements can be seen from Fig. 4. It is essentially a modification of the Schofield-Grace thermocouple.[6]

When first made up the thermocouple consists of four feet each of Pt. and Pt. Rh. wires silver soldered to compensating leads about three feet of which are stripped of their covering and threaded through twin bored fireclay insulators. The rare-metal wires are both covered throughout their lengths with fine, transparent, fused-silica tubing and in addition, excepting the last eight inches or so from the hot junction, are enclosed in one length of ordinary hard glass tubing. This tube is plugged at each end with ignited asbestos to protect the rare-metal wires from contamination as much as possible.

The thermocouple and part of the compensating leads are enclosed in $\frac{5}{8}$ in. dia. thin-walled steel tube of suitable length which is bent over at one end or may have a detachable 90° bend affixed. Whichever method is adopted the bend should be of sufficiently large radius to allow the thermocouple wires and insulation to pass round the bend easily. With this arrangement the thermocouple can be easily withdrawn at any time for examination and replaced in a few minutes. The steel tube is covered with a Morgan graphite plunger-mixer as used for non-ferrous metal mixing. These sleeves have an enlarged end which is convenient for providing a substantial covering for the bend in the steel tube. The portion of the steel tube which has to be immersed in the steel is protected by a block of the same graphite material, or electrode graphite may be used, see Fig. 5. This may be screwed or wedged on to the short verticle

6 F. H. Schofield and A. Grace. A "Quick-Immersion" Thermocouple for Measuring the Temperature of Liquid Steel both Before and After being Tapped from the Furnace. Second Report of the Liquid Steel Temperature Sub-Committee. Eighth Report on the Heterogeneity of Steel Ingots.

7 Third Report of the Liquid Steel Temperature Sub-Committee of the Committee on the Heterogeneity of Steel Ingots. J.I.S.I. 1942. No. I, vol. 145, p. 213.

8 D. C. Gall. A Direct Current Amplifier and its Application to Industrial Measurements and Control. J.I.E.E., 1942, vol. 89, p. 434.

end of the steel tube. This portion of the tube is also lined with ¼ in. bore, thick-walled, fused-silica tube fastened in with Purimachos cement.

The thermocouple is pushed into the steel tube until 2 in. protrudes at the hot-junction end, and this is covered with a thin transparent-silica sheath cemented into the inner lining (fused-silica) of the end piece. The whole requires very little maintenance. The silica sheath can be renewed in a few minutes and one set of the graphite sleeves will stand up to a few hundred temperature determinations. The short length of graphite which enters the slag will withstand 50 to 60 immersions if made of electrode graphite.

The couple complete weighs about 18 lbs. and is easy to handle. It is important in using the apparatus that the thermocouple should be lowered into and raised from the bath with a vertical movement. If an attempt is made to move the thermocouple in the bath, or to drag it from the metal by a combination of horizontal and vertical movement, the resistance of the metal will bend the transparent fused silica sheathed end, which makes it necessary in most instances for the couple to be remade as the insulators may be broken when the silica sheath is removed. With careful use two readings may be obtained from one transparent fused silica sheath and twenty readings may be obtained from one hot junction before it is necessary to remake this. When doing this, half an inch of the platinum couple is cut away before refusing the hot junction. The rejected portion of wire constitutes valuable scrap.

It is perhaps hardly necessary to add that conditions for determining the temperature in the bath must be carefully standardised. Once the most satisfactory methods as regards stirring of the metal before the immersion of the thermocouple, the position in the furnace in which the thermocouple is placed, and other factors, more or less governed by domestic conditions, have been determined they should be rigidly adhered to. It is not convenient or economical to take a large number of readings on one heat, but, should the temperature in the first instance be too low it is possible after a little experience to estimate the length of time and the rate of power input which it is necessary to apply before the steel will have attained approximately the desired teeming temperature. By this method it is possible to approach consistently to within 10°C of the desired teeming temperature.

(*To be continued*)

War-Time Consumption of Non-Ferrous Metals

(*Continued from page* 74)

Cadmium

U.K. Consumption by Main Trades—in Long Tons.

The following table shows the consumption of Cadmium Metal in the U.K. by main trades in 1942, 1943, 1944 and the first quarter of 1945. Figures for the second quarter of 1945 will be published as soon as they are compiled.

Main Trades	U.K. Consumption of Cadmium Metal.			
	1942	1943	1944	1945 1st Qtr.
Plating	109	83	104	48
Alloys	204	142	131	26
Batteries	96	119	109	28
Miscellaneous	21	25	33	12
Total All Trades	430	369	377	114

Cobalt

U.K. Consumption—in Long Tons.

The following table shows the consumption of Cobalt in the U.K. by main forms in 1942, 1943, 1944 and the first quarter of 1945. Figures for the second quarter of 1945 will be published as soon as they are compiled.

Form of Cobalt Consumed	U.K. Consumption of Cobalt (all in terms of Cobalt element).			
	1942	1943	1944	1945 1st Qtr.
Cobalt Alloy	69	113	98	16
Cobalt Metal	314	382	471	118
Cobalt Oxides and Salts	125	131	218	61
Total all forms	508	626	787	195

Manganese Metal

U.K. Consumption—in Long Tons.

The following table shows the consumption of Manganese Metal (including Manganese Content of Master Alloys) in the U.K. by main uses in 1942, 1943, 1944 and the first quarter of 1945. Figures for the second quarter of 1945 will be published as soon as they are compiled.

Main Uses.	U.K. Consumption of Manganese Metal (including Manganese Content of Master Alloys).			
	1942	1943	1944	1945 1st Qtr.
Light Alloys	653	767	546	82
Copper Alloys	535	538	315	62
Total all uses	1,188	1,305	861	144

Summary Table

Consumption of Virgin Metal Only (tons).

	1940*	1941*	1942	1943	1944	1st Qtr. 1945
Copper	440,000	445,000	490,772	448,631	348,139	81,103
Zinc	260,000	256,000	258,564	222,785	184,241	45,411
Lead	250,000	215,000	229,462	197,053	205,385	51,517
Tin	34,500	34,500	23,478	17,631	18,435	5,949
Nickel	23,500	23,000	21,751	17,218	12,420	2,431
Cadmium	—	—	430	369	377	114
Antimony	—	—	3,918	4,197	4,772	1,265
Cobalt	—	—	508	626	787	195
Manganese Metal	—	—	1,188	1,305	861	144

* Approximate.

DR. N. F. BUDGEN has retired from his position as Managing Director of Light Alloy Products Company (Birmingham) Ltd., by mutual arrangement with International Alloys Ltd., which has acquired the interest in the Company. The reconstituted Board has appointed Mr. G. E. Roberts as General Manager.

THE following officers of the Council of the above Society have been elected for the year 1945-46:—*President:* Sir Frederick Handley Page, C.B.E., F.R.Ae.S. Vice-Presidents: H. Roxbee-Cox, B.Sc., Ph.D., D.I.C., F.R.Ae.S., Sir Oliver Simmonds, M.A., F.R.Ae.S., M.P. They take office on October 1st, 1945.

The Yielding Phenomenon of Metals

Influence of Speed and Loading Conditions

Part VI

By Georges Welter

Professor of Applied Mechanics, Ecole Polytechnique, Montreal.

The yielding phenomenon of metals, of importance in applied mechanics as well as in metallurgy, is not yet sufficiently understood. Besides a theoretical consideration, the various factors which have a pronounced influence on the yielding of metals such as the kind of testing machine with its load-indicating device, as well as the rate of plastic deformation of the specimen, have been investigated. In order to obtain additional results about the loading conditions of the specimen, a new direct loading machine with a special recorder was designed, and the stress-strain diagrams of this machine were compared to those of the rigid straining machine. Furthermore, the transmission of the loads from the machine into the test specimen without harmful secondary effects was studied by means of three types of newly developed universal joints which allow loading the specimen axially.

In order to understand the whole stress-strain diagram, an automatic time chronograph was designed.

Furthermore, the characteristics of the two types of machine, the rigid straining machine and the direct loading machine, were compared to each other during the static loading tests, as well as during the dynamic tests with the machine in normal operation. During these tests an artificial yielding effect was obtained by a relative displacement of the loading faces of the machines. Stress-strain diagrams were recorded under different loading conditions; diagrams showing a most characteristic difference between these two types of machines were obtained.

By means of a special device based on sliding friction between three plates screwed together, the dynamic behaviour during yielding of metals in both types of testing machines was analysed and interesting diagrams were recorded. The rigidity of the straining machine could be gradually decreased by inserting elastically stressed elements between the grips and the frame of the machine. In five different steps the characteristics of the rigid straining machine were gradually made nearly equal to those of the direct loading machine.

Comparative tests with usual extensometers measuring in different places the relative movements of the grips and the loading faces of the machine gave valuable indications about the reliability of these instruments.

The upper and lower yield-points of mild steel were studied in function of the speed between about 400 lb. and 20,000 lb. per min. These tests show that the lower yield-point is much more affected by high deformation speed than the upper yield-point.

Time-controlled stress-strain diagrams show in all loading phases important variations of the loading speed in the rigid straining machine and the straining speed in the direct loading machine. The difference between the two types of machines becomes especially effective by comparing the stress-strain diagram of duralumin test specimens tested shortly after quenching and showing a very characteristic yielding effect.

I—TIME RECORDS OF THE TOTAL STRESS-STRAIN DIAGRAM (executed on rigid and soft testing machines)

Besides the different phases of the yielding period, it would be of some interest to know how the materials behave with regard to their loading and deformation speeds in the elastic range of the stress-strain diagram in the different types of machines. In the usual testing practice, tests are made with an initial definite loading speed in the elastic part of the diagram. This is obtained by regulating the straining speed of the machine. However, not only in the yielding period, but also under further plastic deformations, the loading speed may change appreciably in the different plastic phases of the test. In order to investigate the different speeds in the stress-strain diagram after yielding, some complete diagrams were made with simultaneously recorded time diagrams.

(a) Rigid straining machine

As represented in Fig. 74, the straining speed on the rigid machine was for diagram No. 1 of about 30 secs. for 0·1 in. deformation, for No. 2 it was about 12 secs., and for No. 3 only 4 to 5 secs. for the same plastic deformation of 0·1 in. The diagrams of the time record show, immediately after yielding right in the beginning of phase *d–e*, a load increase of about 20 lb. to 30 lb. per sec.; in point *e* an increase of only about 6 lb. to 7 lb. per sec., falling down to less than 2 lb. to 3 lb. per sec. in *f* near the maximum load. In point *g* no further load increase takes place; here the specimen is exclusively strained as during the whole test with about 0·2 in. per min., and this under nearly constant load. After the maximum load was reached an appreciable drop of the load, due to the excessive local straining of the test specimen, is recorded. Shortly after the maximum load is attained in *h*, the speed of the load decrease is already about as high (20 lb. to 30 lb. per sec.) as the maximum increase in the first plastic phase after yielding, rising up to 50 lb. and even 60 lb. per sec. in the last moment in *i*, just before rupture takes place. It is evident that an appreciable drop of the load in the last phase of the plastic deformation has nothing to do with the mechanical properties of the tested material. It is merely a sign that the force equilibrium in the rigid

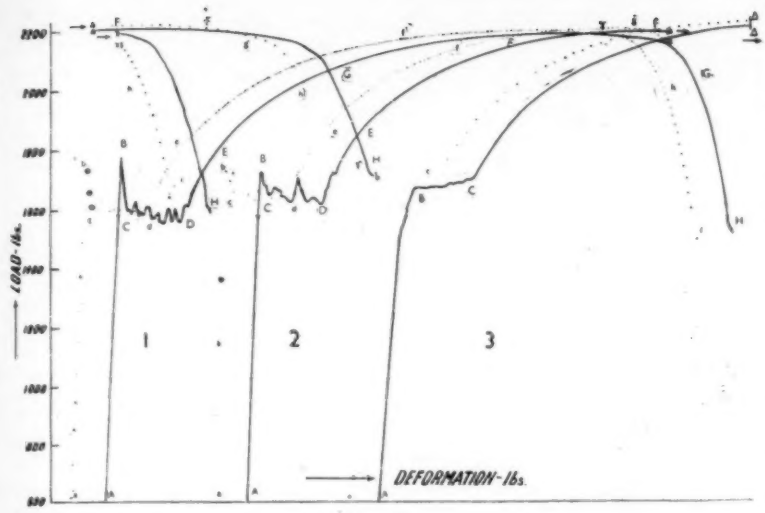


Fig. 74.

straining machine is disturbed by the excessive local straining of the specimen, provoking this drop of the load from 2,186 lb. down to about 1,500 lb. Depending of the degree of rigidity of the machine, this last part of the diagram F–G–H may take, as will be shown later, a completely different shape concerning the load decrease in function of the deformation of the specimen.

Diagrams Nos. 2 and 3 represent similar tests executed with higher deformation speeds of the specimen. Straining took place at 12 secs. per 0·1 in. deformation for diagram No. 2 and as fast as over 1 in. per min. (4 to 5 secs. per 0·1 in. deformation) for diagram No. 3, Fig. 74. For such a high deformation rate, not only the yielding of the material is remarkably influenced, but also the load decreasing phase D–E is different from that of the preceding diagram. The drop of the load is not so steep, the rupture of the specimen takes place somewhat earlier and the load drops in the moment before rupture with a speed of more than 2,000 lb. per sec. This diagram shows also in its elastic phase A–B and in the yielding period B–C that the cord extensometer is not able under these very high speeds to record the true characteristics of the test specimen.

(b) Direct loading machine

This type of machine gives, not only in the region of the yielding of mild steel as already shown, a very characteristic and completely different stress-strain diagram from that of the rigid straining machine, but also in the plastic range of all test specimens, the diagram is distinctly different and of special interest. As this is generally the case in mechanical application and is obtained by the construction of the direct loading machine, no drop of the load takes place after the maximum load is attained and this up to the moment of rupture of the specimen. Beginning from the first yielding of the material the rate of deformation increases regularly. This is in complete contradiction with the results of the rigid straining machine where this rate is, right from the beginning and throughout the whole test, completely constant. Furthermore, at the ultimate load there is no drop of the applied load which remains practically constant up to the very end of the test.

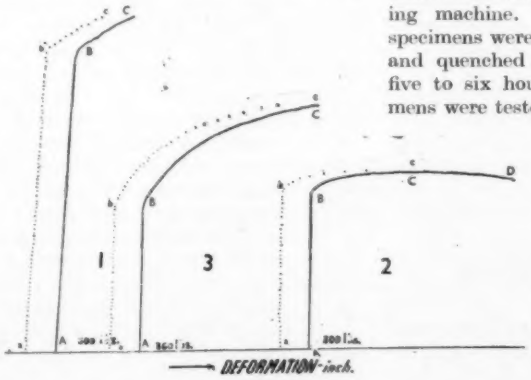


Fig. 75.

In Fig. 75 are represented three stress-strain diagrams with simultaneous time records obtained on the direct loading machine. Diagrams Nos. 1 to 3 are greatly different from those of the rigid straining machine. No. 1 is the diagram of cold worked mild steel. After a loading rate of about 1,400 lb. per min. during a regular increase of the elastic deformation A–B, the load increase from *B* is slower while the plastic deformation becomes regularly greater; between *b*–*c* the distance of the time steps is gradually increasing. Furthermore, after the maximum load in *C* is attained, there is no load decrease recorded. At its highest load the diagram ends at the same moment at which rupture of the specimen takes place. A quite different diagram is obtained from that recorded by the rigid straining machine. Another diagram with yet greater plastic deformations is represented with the time record in No. 2, Fig. 75. The load increase in the elastic phase A–B is about the same as for diagram No. 1. In the plastic range the load increases slowly to a maximum in about 5 to 6 secs. and suddenly after a very great acceleration of the deformation, the specimen breaks in *B* having been strained from *C* to *D* in less than 1 sec. Diagram No. 3 is that of annealed aluminium with its time record showing an analogous result in the direct loading machine as in the previous diagrams. From the elastic limit, the deformation increases progressively in the plastic range of this material till it marks a high deformation rate at a nearly constant maximum load at the rupture of the specimen.

Fig. 76 represents three diagrams of a light alloy tested under the same conditions on the direct loading machine as well as on the rigid straining machine. The duralumin test specimens were heat-treated at 500°C. and quenched in cold water. After five to six hours ageing, these specimens were tested. It is a well-known fact that this alloy shows under increasing loads a few hours after quenching a very characteristic crystal slipping effect, visible in the record of the stress-strain diagram and audible by sharp cracks in the specimen. This slipping taking place in a certain number of crystals or groups of crystals is repeated a number of times during the straining process, the diagram of the rigid machine resembling somewhat the first sliding period of the yielding phenomenon in mild steel. This alloy seems to be of some interest

for comparative tests in the direct loading machine and in the rigid straining machine. As represented in Fig. 76 these two kinds of diagrams are indeed quite characteristic and the results are of remarkable interest.

In the direct loading machine the record No. 1 of duralumin shows a load deformation diagram interrupted in the beginning by numerous small horizontal steps; these steps become larger and fewer with growing loads till the rupture of the specimen suddenly takes place at a maximum load in *C*. The time record indicates that these horizontal steps appear instantaneously in the diagram; there is no time mark for most of the bigger steps between their beginning and their end. The time record proves that these steps appeared in less than 1 sec., and by direct observations it may be said that they occurred instantaneously, that means in far less than $\frac{1}{10}$ sec.

A quite different picture of this material was obtained in the rigid straining machine as shown in diagrams No. 2 and 3, Fig. 76. Here in no case was a horizontal step recorded, and the sliding effect became visible in the diagram by a steep vertical drop of the load for each single slip. Here too this effect is not very pronounced in the beginning of plastic deformation, but with growing loads it becomes more and more visible. For diagram No. 2 the maximum drop of the load due to slipping in the specimen is of about 300 lb., while for diagram No. 3 the maximum value is as high as 500 lb. to 600 lb. In this diagram No. 3 only five very pronounced vertical steps of this kind took place, while diagrams No. 1 and 2 showed a much greater number of these effects.

Here again the very important and characteristic difference between the direct loading and the rigid straining machines can be observed, a difference which can no longer be ignored in modern testing practice of metals.

J—GENERAL CONSIDERATIONS

Considering the results of the preceding investigation, it may be said that the direct loading machine has many advantages over the usual rigid straining machine. The direct loading machine reproduces much more faithfully the conditions obtaining in practice. In fact the applied loads in most of the constructions act in such a way that the stressed element is not discharged by the yielding process. The load remains the same even if through

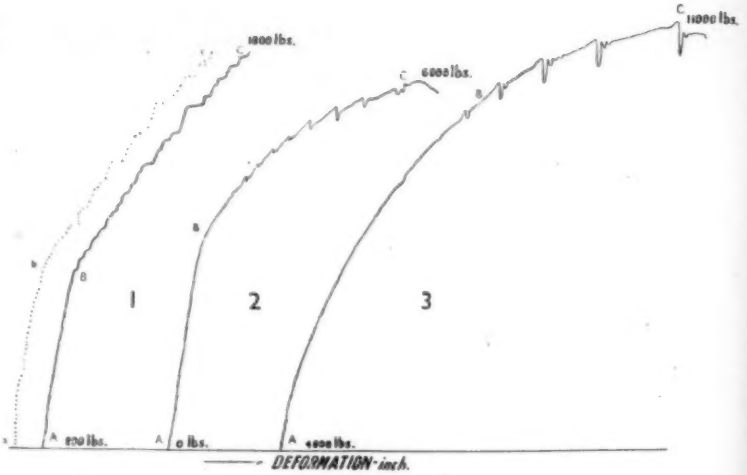


Fig. 76.

overloads the metal deforms appreciably by stretching. It seems rather important to follow as closely as possible the real behaviour of the solicitations in practical applications of metals; here the direct loading machine has, amongst others, a very great advantage over the rigid straining machine. Unaffected if the specimen strains or yields, the direct loading machine keeps practically a constant load; in the rigid machine, however, the load provoking strain falls instantaneously down to very small values, interrupting in this way the important deformation period of the specimen under test. The amount of the drop of the load depends on the degree of stiffness of the machine and can vary between very wide limits. This sudden unintended interruption of the yielding process in the rigid straining machine can be considered as undesirable. The critical and important phase of the transition from elastic to plastic behaviour of the material showing discontinuities in the stress-strain diagrams, is right at the beginning interrupted by the considerable self-discharging effect of this kind of machine, causing very complicated and nearly uncontrollable reactions in the specimen itself as well as in the load indicating device of the machine. It is evident that this behaviour of the rigid straining machine cannot be credited as an advantage of this type of machine. All is well as long as the deformation is in some accordance with the load increase. If, however, a sudden additional deformation in the specimen under test takes place, then the whole equilibrium of the rigid straining machine is completely disturbed and complicated reactions in both the machine and the specimen are the result generally classified as the yielding phenomenon of metals. Instantaneous changes in the specimen by sudden longitudinal axial deformation provoke on the rigid straining machine a kind of catastrophe in the records registered by the machine. In an extremely short time the specimen changes over from elastic to plastic deformation while the testing machine itself cannot follow this rapid movement; an uncontrollable fairly delayed drop of the load as well as an unreliable record of these small deformations are obtained. Independently of the extensometers, or the kind of direct observation used to measure the micro-deformations of the initial period of the yielding phase, no record of a rigid straining machine can give a true picture of this phenomenon. We can see that, based on the results of the above investigation, the observations about the straining speed in the dropping period of the load during the discontinuities in the stress-strain diagrams are not reliable. They do not give any information about the real behaviour of the test specimen.

Besides this, the advantages of the soft direct loading machine, as compared to the rigid straining machine, may be enumerated as follows: investigations under constant loads can be made and no dropping takes place at the ultimate load; rapid plastic deformation of the specimen can be truly recorded as well as every single point of the yielding period; the weighing system of the machine remains in equilibrium during the whole testing period because no abrupt change of the load takes place. By inserting springs between the frame

and the grips of the rigid machine, the springiness of the machine can be highly improved; elastic elongation and even the elongation during yielding of the test specimen are, in this case, relatively small compared to the elastic deformation of the machine frame.

In connection with the above statement, the study of this subject is not only of interest for the design problems of elements as well as for the metallurgist, as has been said at the beginning of this work, but another more important question must be pointed out. We know that in the materials testing technique the phenomenon of the dropping of the load during yielding is of basic importance in connection with the stress-strain diagrams of different metal specimens. Besides the usual type of tensile stresses to which structural members as well as machine members are subjected, the very important group of compressive stresses may be mentioned. For compressive members subjected to end-loads the stresses during testing are generally recorded in function of the strains. The recorded stress-strain diagram, for columns, for instance, consists in a straight line showing an elastic deformation up to the maximum load where collapse takes place. If, however, other structural elements are tested under compressive loads, as for instance, thin sheet elements used in airplane construction, then the stress-strain diagrams have a quite different shape; they resemble in many points to the tensile diagram of mild steel showing the distinct dropping of the load at critical stresses. In Fig. 77 are represented a few characteristic diagrams recorded during buckling tests of curved aluminium alloy sheets under compression stresses, executed in the same rigid machine used for the present investigation. These diagrams show after the elastic phase *A–B* the characteristic drop of the load from a critical maximum load *B* to a minimum load *C*. This drop of the load is accompanied by a very sharp and audible crack while the first buckling effect in the plate can be observed. After this first drop *B–C* the load continues with a greater slope to increase again up to a second critical load *D* where another smaller or even more important drop of the load takes place. It is evident that these records on which are based most of the designs in stressed skin members of airplanes, are not in complete accordance with the true behaviour of these elements in practice. If the critical load in *B* is attained and buckling takes place, then right in this moment the applied critical load is in practice not reduced instantaneously to an appreciable smaller one in point *C*. This drop of the load is merely an effect of the rigid straining machine. If the stressed platens in an airplane wing for instance are well designed and adjusted with precision they should theoretically support, all or a group of them, a maximum load at the same moment. The maximum load, however, being in extreme cases not limited to the critical load in *B*, continues to act during buckling and is not reduced to a smaller value as this is the case for the buckling tests in the rigid straining machine. It seems that, based on the preceding investigation, it would be more logical and the results should be more reliable if this kind of test was executed on direct loading machines. The application of the load in this type of machine seems to be in accordance with what really happens during the buckling effect in practice. It seems to have become necessary to make a revision of the results obtained on the rigid straining machine which shows the unexplainable dropping of the load in the very critical and important moment of the buckling effect. Results obtained during buckling tests with thin aluminium alloy sheets in compression executed in the direct loading machine should give a more reliable understanding and further valuable indications about this interesting and complicated phenomenon.

K—RESUME OF RESULTS.

The results from the above investigation may be summed up as follows:

(1) The shape of the stress-strain diagram of metal test specimens depends to a very high degree on the type of testing machine.

(2) The yielding phenomenon of metals, as recorded by the rigid straining machine, is not in accordance with the true behaviour of the test specimen.

(3) The sudden decrease of the load at the beginning of the yielding period is an effect of the rigid straining machine and is not recorded simultaneously with the yielding process itself.

(4) In the direct loading machine no upper and lower yield points are recorded. The total yielding takes place instantaneously at a single critical load. This load corresponds to the maximum load in the elastic phase.

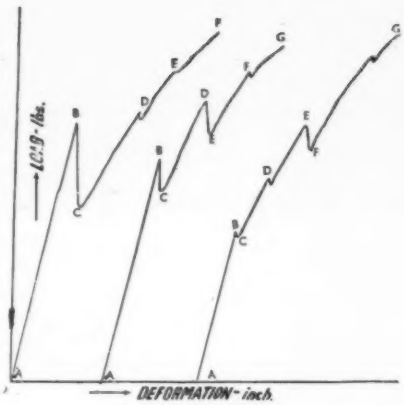


Fig. 77.

(5) The test specimen breaks in the direct loading machine, as this ought to be for all machines, at the ultimate load and not at an appreciably lower one.

(6) The rigidity of a usual straining machine has been gradually reduced in five different steps and the stress-strain diagrams recorded.

(7) The rigid straining machine can be transformed into a direct loading machine with practically the same loading and straining characteristics.

(8) The modulus of the rigid straining machine has a marked influence on the results recorded during yielding.

(9) Time-controlled stress-strain records of the total test period were made on the rigid machine and compared with those of the soft machine.

(10) If the specimen elongates plastically in the rigid machine, it provokes a discharge of the elastically strained frame of the machine followed by an appreciable load decrease. The specimen elongates right into the stressed machine frame without any perceptible movement of the loading tables.

(11) The usual cord recorder attached on the cross-heads of the machine registers their movements but not the straining of the test piece itself during yielding.

(12) The electrical extensometer adjusted on the test specimen gives a true record of the strain in function of the stress and time during the elastic period; during the initial yielding period, however, the record is in accordance with neither the real drop of the applied load nor the elongation of the specimen.

(13) For very fast yielding movement approaching dynamic testing conditions the usual extensometers are completely unreliable.

(14) With an increasing rate of loading, the slope of the drop of the load in the diagram of the rigid machine becomes less steep having a tendency towards a horizontal part during the initial yielding period.

(15) The rate of straining influences the lower yield point more than the upper one; up to 20,000 lb. per min. the lower yield point is increased by about 16%. Already for this high deformation speed the usual extensometer does not give a true picture of this value.

(16) A new tensile testing machine based on the principle of direct loading by a dead weight and a lever arm as well as a stress-strain recorder were designed, constructed and calibrated. This machine gives completely reliable results during the yielding period as well as at the breaking load of the specimen.

(17) A close time control of the different yielding phases was made possible by a newly devised automatic chronograph, registering the load and strain increments automatically and simultaneously with the stress-strain diagram in function of the time.

(18) Valuable information about the different phases of yielding and the breaking process have been obtained by the automatic chronograph.

(19) A new sliding device, imitating fairly naturally the yielding process of mild steel or the sliding process of single crystals, was originated. With this device, useful results during the yielding process were obtained.

(20) By means of three different types of universal joints it was tried to obtain axial loading conditions, absolutely necessary for a reliable study of the yielding phenomenon.

(21) An artificial stress-strain diagram was recorded on the rigid straining machine but without test specimen, by changing the relative movement between the loading tables by a special device.

(22) The present results about the yielding phenomenon may give reliable information for tackling the most important problem of thin curved light alloy sheets in compression (buckling tests). A closer verification of the results obtained on the rigid straining machine, by using a direct loading machine, may furnish a quite new aspect to this problem, which is of current interest in stressed-skin airplane structures.

L—RECOMMENDATIONS

Based upon the above test results as well as the developed testing apparatus, an investigation of the following problems directly related to the subject of the yielding phenomenon of metals, seems to be of basic importance in the design of steel structures and machine members as well as in the enlargement of our physico-metallurgical knowledge about metals:

(1) Development of recorder devices giving absolutely reliable results during the suddenly interrupted stress-strain process due to rapidly changing effects during the yielding period.

(2) Effect of the modulus of the existing testing machines on the testing results with propositions for standardisation of all testing machines in this regard.

(3) Development of automatic time recorder devices adaptable to the usual testing machines.

(4) Universal testing machines with a wide range of variations regarding their factor of rigidity, that is a range of possibilities to execute tests on the same machine under soft and rigid conditions.

(5) Control of the more important results of buckling tests made on rigid straining machines. Useful and highly interesting results for aeronautical design would be obtained by these tests using direct loading conditions for the study of these important problems.

(*Concluded.*)

Corrosion of Yellow Brass Pipes

By E. H. Polushkin and H. L. Shuldener

THE corrosion of yellow brass pipes removed from domestic hot-water installations was studied under the microscope in relation to their microstructure and condition of service. The pipes which were made by different manufacturers and whose sizes ranged from ¾ in. to 4 in. nominal inside diameter, were divided into two groups according to their composition: (1) With a copper content of 60–62% (22 pipes) and (2) with a copper content by 65–68% (18 pipes). In examining the pipes particular attention was given to the study of the structural aspects of corrosion, the origin of local corrosion, the nature of copper deposits, and general conditions of dezincification.

All pipes were found to be corroded, but the rate of general corrosion per year was small, and some specimens were in good condition after 20 to 25 years' service. Most pipes, however, exhibited local corrosion which was much more destructive. Grain size had apparently no effect on corrosion, except perhaps, when it varied considerably, but the surface defects of pipes, the residual straining of metal and cracks were influential factors. Corrosion of both constituents of brass, alpha and beta, had two distinct stages, first dezincification and second the final corrosion of copper formed, and the lag between these stages varied considerably. In alpha-beta brass pipes, the beta was always attacked first, but here also the lag varied considerably.

Microscopic examination also revealed that local corrosion in some pipes was closely associated with residual stresses and local deformations, and the evidence thus obtained pointed to the conclusion that these factors were influential in starting the localised corrosion in these pipes. Conclusive evidence of the residual origin of copper derived from dezincification of the alpha constituent was found in the structural characteristics of the deposits. In regard to copper derived from dezincification of the beta constituent, the residual origin is the most probable.

In the examination of the pipes, the presence of alternating zones in plugs and zonal structures in tubercles, shown by the presence of many concentric shells in the walls, indicated that some periodic changes were involved in the process of corrosion. Further evidence of periodicity was also found in corroded "beta," which displayed parallel bands of copper and cuprous oxide arranged in alternate order.

Metals Technology, 1944, vol. 11, No. 7, and A.I.M.M.E. Technical Publication No. 1742, pp. 17.

ANY doubts which may have existed as to the adequacy of the response to the policy promised by the Microchemistry Group must certainly have been set at rest by the success of the Manchester meeting, reported elsewhere in this issue. The policy, one of the principal aims of the group, of regarding regional centres as equally worthy of attention with London, and therefore equally fit to act as important meeting places, was inaugurated by this, the second meeting of the Group. The audience was large. More than this, it was representative ; members of the Group had attended from widely separated parts of the country to listen as well as to speak. Most important, it was interested. Lively discussions after each of the three papers were a sufficient indication of the importance attached to the spread of microchemical information. A special word of praise must be added for the efficient and valuable demonstration arranged for the benefit of the Group by Professor Kenner at the College of Technology. As we indicate elsewhere, this demonstration, quite apart from the interest centred in the methods, was of prime importance in that it indicated clearly what is possible in the way of teaching microchemistry under difficulties. Other centres which are to be visited by the Group have something to which to look forward, if they are lucky enough to receive such stimulating fare as was provided on this—we will risk being trite, and say it !— this quite literally auspicious occasion.

Problems Concerning the Microchemical Balance—Part II Cleaning

By David W. Wilson

The microchemical balance is one of the most sensitive and delicate, and at the same time one of the most fundamental instruments employed in analysis. Fairly frequent cleaning is necessary, but it can easily be rendered useless, and indeed detrimental, by damage inadvertently caused during the process. The equipment and precautions necessary for the cleaning are described.

TO the question, "How often should a microchemical balance be cleaned?" there is no universal answer, unless it be, "As seldom as possible." The period between cleanings will depend on the efficiency of the provisions for the exclusion of dust, as well as the frequency with which the balance is used. If the balance-room is air-conditioned, or in a locality where dust is at a minimum, the balance may function satisfactorily during six months of daily use. With a less fortunate situation cleaning may be necessary as often as once every few weeks. A double balance-case (the outer one is conveniently a converted fume-cupboard which has been made reasonably dust-proof) will help to diminish the number of cleanings required, but the frequency of cleaning can only be determined by the experience of the user.

The balance will, of course, undergo careful cleaning when it is first assembled, and the following recommendations are intended to augment the suppliers' instructions to those installing a microchemical balance, as well asto act as a guide to those undertaking its routine overhaul.

The first indication that a balance requires attention is generally a tendency for the swings to become erratic, due to tiny fibre particles settling on surfaces where moving and fixed parts are close together. Failure to obtain repeatable weighings, which cannot be explained on the grounds of electrostatic charges or variations in temperature or humidity, may be due to the same cause. When this occurs, close observation through a lens should be made of the pointer tip and the central knife-edge. A single hair on the pointer or pointer scale can upset the swinging of the beam, and if the fault lies here no further cleaning may be necessary than a gentle brushing of the part with a small camel-hair brush. But if, on examination of the knife-edge by light reflected from an electric torch, fibres or dust are observed, a general cleaning is advisable.

A start is made with the surroundings of the balance. The balance bench and, if there is an outer case its interior ledges, are thoroughly dusted. A small glass attachment to a vacuum cleaner or vacuum line will remove dust from awkward corners without stirring it

up. Any dust which is disturbed should be allowed to settle before opening the balance-case.

Some balances have provision for the removal of the complete case. If this is possible, it should now be done, and the case removed from the balance-room for cleaning and polishing. If only the front window can be removed, the outside of the case should first be cleaned (the vacuum attachment is useful for the ledges) and then the window lifted out and laid, after its inside surface has been cleaned, on the top of the case. It will generally be impossible to detach the window from its sash-cords.

A clean sheet of plate-glass (or, less satisfactorily, glazed paper) is laid down at each side of the balance. On a third sheet in front of the balance are assembled the cleaning materials. It is hardly necessary to point out that these should be kept in a close-fitting box or drawer when not in use, and used only for the one purpose. It is advisable to wash one's hands at this stage, and it may be deemed necessary again before reassembling the parts. In some ways, and with a certain amount of justification, cleaning a microchemical balance is comparable with a surgical operation. The enemy here is not infection, but perspiration, and perspiration can corrode the metal parts of a balance as surely as germs can attack the membranes of the body. Corrosion, once started, develops quickly, and a corroded balance is neither pleasing nor apt to be accurate.

Some protection, then, for the fingers is indispensable and chamois is excellent for the purpose. A set of short finger-stalls can be cut from a discarded chamois glove, or made from squares of chamois, and give more freedom of movement than a glove. Four, one each for the thumb and first three fingers of the right hand, are sufficient and are easily adjusted. Two camel-hair brushes, one large and flat and the other small and pointed, are used for removing dust, and a few squares of chamois will be needed for cleaning the metal and glass surfaces. A chamois square wrapped round a pointed match-stick is used for cleaning out the supporting cups of the beam and stirrups, and a supply of the match-sticks should be to hand. They should not come to too fragile a point, and can be smoothed off on the abrasive side of a match-box, if care is taken that no particles of abrasive remain embedded in the wood. The ivory-tipped forceps from the box of weights will be required ; an electric torch and magnifying lens (the latter to show up dust by reflected light from the former) and a tube of light oil are also necessary. Any light machine oil will do, but watch oil is to be preferred, and should be kept apart from the other cleaning materials. Some ordinary tissue paper is useful for cleaning oil and dust from the moving parts under the base, and a book of special tissue paper (the grade supplied for cleaning microscope lenses) is used to wipe the knife-edges.

When these materials have been assembled, the rider is removed with the forceps and placed for safe keeping in the box of weights. The rider-lens and pointer-lens or telescope are detached and placed to one side of the balance-case.* Before the pans and stirrups are removed the markings on them should be studied. They will be marked, possibly corresponding to similar marks on the ends of the beam, either with the letters R and L or with one dot for left-hand side and two dots for right-hand side. The dots on the pans and stirrups should be towards the front. When it is certain that they can be replaced properly the pans are removed and placed on the glass, each to its own side of the balance-case. The stirrups are placed in front of their respective pans.

The beam is next removed. This is an operation calling for a steady hand and an ability to watch simultaneously the two most sensitive points—the pointer tip and the central knife-edge. It is held, still, of course, with finger-stalls, by the upper part of the pointer, lifted clear of the supporting cups, and the pointer tilted clear of the pointer scale. If the tip is behind the scale and is observed by reflection from a mirror (as it is in some microchemical balances) this operation requires even greater care. The central knife-edge is meanwhile kept clear of the agate plate and its supports, and lifted off. It may be that the beam cannot be removed until the arresting mechanism has been lowered ; in this case the weight of the beam is taken by the fingers, then the mechanism lowered, using the left hand, care being taken that the central knife-edge does not touch its agate plate while the beam is tilted and lifted clear. The beam is laid on the glass plate, with the pointer tip slightly elevated, where it will rest with the tip clear of the plate. The finger-stalls can now be removed.

The interior of the case now receives a careful cleaning, first with the large camel-hair brush and then with a chamois square ; the rider hook and its mechanism are not forgotten. It is inadvisable to oil any parts inside the balance-case. On no account should the screws securing the central pillar to the floor of the case be loosened or removed, as a slight displacement of the pillar may cause the arresting mechanism to stick, as well as throwing the beam out of alignment with the rider hook and pan arrests. The pillar and the beam support are brushed with the small brush, and then wiped with a chamois square. The two short vertical pillars on which the beam support slides should be wiped as clean as possible, although it is often very difficult to reach them. The pan arrests can generally be lifted out of the floor of the case, and wiped clean (not oiled) and placed on the glass plates at their respective sides of the case.

A piece of chamois wrapped around a match-stick is used to wipe out the cups supporting the beam and the stirrups. The central agate plate is next cleaned. Using the finger-stalls, a square of special lens tissue is folded into a spill and then in half, so that the fold can be rubbed lightly over the agate surface. Then, after removing the finger-stalls, all the surfaces are examined by reflected light from the torch and any remaining particles of dust or fibre removed with the small brush.

Attention is now turned to the rotating and sliding parts under the floor of the case. These are cleaned (a piece of ordinary tissue is best for this) and oiled *very lightly* (most people realise that too much oil merely collects dust, but most people continue to apply too much oil) using the point of a match-stick. The important parts are the bearings of the axle which operates the release, and the cams and axles attached to this movement. If the pan arrests make sliding contact with a lever underneath, the point of contact receives attention. Oiling may not be necessary each time the balance is cleaned, and if it is omitted, re-levelling will probably

* It is sometimes recommended that the parts of the balance, on removal, should be placed on the glass window, which is now on top of the balance-case. This is in general permissible, but with narrower balance-cases it may, through instability, expose the delicate parts to unnecessary risks. The alternative procedure described here is then markedly superior.

be unnecessary. After the pan arrests have been replaced, the handle is turned to ensure that the mechanism works smoothly, and the level of the balance readjusted approximately, using the plumb-line or spirit-level.

Before cleaning and re-assembling the beam and pans, the hands should be washed again to remove traces of oil. Then, using the finger-stalls, the beam is held by the upper part of the pointer and brushed over with the small brush, and the supporting points wiped with chamois. The pointer tip should receive special attention. Folded squares of special tissue are used to wipe the three knife-edges, after which careful inspection is made of all surfaces to ensure their freedom from dust. The beam is then replaced on its supports. Careful watch is kept of the pointer tip and the central knife-edge and, reversing the previous procedure, it may be necessary first to lower the arrest mechanism and hold the beam in position clear of the central agate plate while the mechanism is raised to support it. This operation requires even more care than the removal of the beam.

The stirrups are cleaned by brushing and wiping the supporting points with chamois, and the agate plates with a folded square of special tissue, and they are replaced, in their correct positions, on their supporting cups. A chamois square is used to polish the pans. If by some catastrophe or gross carelessness the pans have become stained or corroded, they can be cleaned with a little metal polish, but all trace of polish must be removed with a soft cloth followed by chamois. If polish is used too often, the plating is bound to suffer. The tips of the pan arches and the hooks for holding tubes are given a careful cleaning, and after inspection for absence of dust they are replaced on their respective hooks. The arches must be guarded against deformation which would affect the pan arrests and result either in the pans not being supported while at rest or oscillating when released.

The rider is held by the forceps, brushed with the small brush, and replaced on its hook, and the rider lens is polished, replaced and adjusted into focus. The pointer lens or telescope is similarly cleaned and adjusted, and the balance-case or front window replaced. Final levelling is carried out, and the mechanism is tested to ensure that it works smoothly.

It will be necessary to allow the balance to "settle down" that is, to allow the temperature of the inside of the case to become uniform. This may take up to an hour, during which the window and side doors are left open. Then the zero is checked and adjusted, if necessary, by turning the adjusting screw with the forceps.

While the balance is settling down, the accessories are cleaned. The weights are removed from their box on to a sheet of glazed paper, and the box is cleaned. The vacuum attachment is used for the velvet lining, and any remaining fibres removed with a stiff brush. The weights are cleaned with chamois and replaced (occasionally they may be washed with a 50/50 mixture of pure alcohol and distilled water and dried thoroughly). Tares are removed from their drawer, the drawer cleaned and the tares wiped with chamois and replaced.

Finally the cleaning materials are put away, with the exception of one large chamois square; the glass plates are removed, and after closing the windows of the balance-case the surroundings are wiped, the windows of the outer case, if there is one, polished, and the electric light bulb or fluorescent tube cleaned.

The microchemical balance is now ready for use again. Cleaning has taken some time, during which both the balance and the operator are not "productively" employed. This may be used as an argument against unnecessary "tinkering" with the balance, but should never be regarded as wasted when the balance requires attention. Failure to keep the microchemical balance in good condition leads to its early deterioration and destroys the natural pride of its operator in one of the most satisfactory tools of analysis.

The Microchemistry Group

THE first visit to a provincial centre of the Microchemistry Group of the Society of Public Analysts and other Analytical Chemists took place in Manchester on Friday, 25th May. The meeting was held in conjunction with the local branch of the Royal Institute of Chemistry.

As an introduction to the meeting, a demonstration of apparatus and methods as used in the College of Technology was introduced by Professor J. Kenner at the College. Professor Kenner pointed out that partly through natural disadvantages of situation, and partly because of the war, the College was by no means ideally equipped for the practice of microchemistry. However, the belief which he had in the advantages of some acquaintance with the methods had caused him to introduce, quite some years ago, a selection of the methods into the ordinary courses. This would be extended as time allowed and opportunity arose. The primary aim of the demonstration was to show that any Institution, under any conditions, could make a very adequate attempt to teach some at least of the methods.

A special balance for delicate weighings, a demonstration of semi-micro and drop analysis after the methods of van Nieuwenburg, and a selection of the more important organic quantitative methods used by the College on the semi-micro scale, were then open to the visitors, and were described by members of the staff.

From the demonstration, members of the Group passed to the Grand Hotel for the reading of papers. This section of the meeting was opened by a short address from Professor H. V. A. Briscoe, chairman of the Group, who presided. Professor Briscoe reviewed the policy of the Group, and described how the committee was advancing the welfare of microchemistry in other ways as well as by holding meetings. He instanced an investigation into the teaching of microchemistry throughout the country, which was, in turn, giving a basis for further action.

The first paper, on "Some Physicochemical Methods in Microchemistry", was given by Dr. Cecil L. Wilson. Dr. Wilson pointed out that the popularity of some microchemical methods tended to overshadow others which were equally ingenious and extremely useful. Physicochemical methods (as distinct from purely instrumental methods) were one branch which had thus suffered somewhat from neglect, and many of the methods were not so well known as they should be. Yet some study

of them should go far towards convincing one that nothing was impossible on the micro scale. Dr. Wilson briefly reviewed representative methods which had been proposed for the determination of viscosity, surface tension and refractive index, some of the methods paralleling macro methods closely, but others using quite novel principles.

Mr. G. Ingram spoke on the "Determination of Trace Amounts of Sulphur Dioxide", with special reference to the microdetermination of sulphur in steels. A review of existing methods indicated the flaws which prevented any of these from being completely satisfactory. Mr. Ingram had devised a method by which the steel is burnt in oxygen, and the resulting sulphur dioxide collected in hydrogen peroxide in a special absorber. The liquid is washed from the absorber into a platinum crucible with conductivity water, and treated with copper oxide. Dissolved copper is estimated absorptiometrically by sodium diethyldithiocarbamate, and corrected by a blank determination. A range of 49 to 4·9 μg. of sulphuric acid can be estimated in steels with an accuracy of less than $\pm$ 3% recovery of sulphur. The method can also be applied to organic compounds, although the presence of nitrogen causes complications which have not as yet been overcome.

Miss M. Correr, dealing with micro methods used in the analysis of cotton, pointed out that in the course of its treatment cotton comes into contact with many materials and varied processes. Methods have to be devised which will detect or determine impurities or faults which have arisen as a result of this treatment. Thus, chemical "tendering" can often be indicated adequately by a process of silver staining. Qualitative tests for metallic impurities, by spotting or staining on the actual fibre, may often give sufficient information. Such elements as copper, manganese and iron are often dealt with in this way. Quantitative estimations involve the special problems associated with the determination of very small amounts, frequently under 1 mgm., present in large samples which often range from 1 to 5 g. Metals such as the above are usually determined by specially modified variations of standard methods. The estimation of antiseptics applied to fibres is a problem which often arises, and for which special methods have been devised.

Polarographic Determination of Magnesium

MAGNESIUM is unsuited to direct polarographic determination,[1] but this difficulty has been overcome by precipitation with 8-hydroxyquinoline, filtering off the hydroxyquinolate, dissolving this in hydrochloric acid, and measuring the liberated 8-hydroxyquinoline polarographically, this, of course, being equivalent to the original amount of magnesium.[2] It is inevitable that there should be difficulties in the filtration and handling of the magnesium-8-hydroxyquinolate, and a recent method[3] endeavours to avoid this.

Five mls. of a standard 8-hydroxyquinoline solution (a solution containing 0·5 g. per litre of 5% ethanol is suitable) and ten mls. of a buffer solution are placed in each of two 25-ml. graduated flasks. The buffer is 0·036 M in ammonium chloride, and approximately 0·25 M in total ammonia, the pH being adjusted to 10. One of the 25-ml. flasks is made to the mark with distilled water, while to the other is added a given amount of the magnesium solution, and then this, too, is made to the mark. Both solutions are mixed thoroughly, and then polarographed. The decrease in wave height of the 8-hydroxyquinoline is found to be directly proportional to the amount of added magnesium between the amounts 5 and 200 μg. magnesium. The slight solubility of magnesium-8-hydroxyquinolate may give rise to an error of not more than 1 μg. of magnesium per 2·5 mls. of polarographed solution.

Any ion precipitated by 8-hydroxyquinoline at pH 10 will normally interfere, but all of these, except titanium, may either be simply removed, or have the effect nullified by, *e.g.*, conversion to a complex.

The method would appear to be more convenient and rapid than that of Carruthers, and it has been successfully applied to such low concentrations of magnesium as are found in tap water, *i.e.*, 5 parts per million, or to the determination of magnesium in plant ash.

1. J. T. Stock. *Metallurgia*, 1944, **29**, 155.
2. C. Carruthers. *Ind. Eng. Chem., Anal. Ed.*, 1945, **15**, 412.
3. K. G. Stone and N. H. Furman. *Ibid.*, 596.

Spot Tests Applied to Steels

RECENTLY, in an investigation of spot tests for distinguishing certain classes of steels, it was found possible[1] to draw up a scheme including practically all the elements likely to occur in steels. By this scheme, a series of tests can be applied which are in general sensitive to at least 0·1%, are reliable, and are not interfered with by the other elements which one is likely to find present.

The elements detectable in this way are nickel, chromium, manganese, molybdenum, tungsten, aluminium, copper, lead, titanium, cobalt, selenium, vanadium. Treatment for the avoidance of interference is usually simple, and most of the tests are carried out merely by stripping a small spot on the surface of the steel and transferring this to spot paper or a spot plate with the appropriate reagents. The only special apparatus required consists of small capillary pipettes and home-made filters devised from 10-mm. lengths of 8-mm. diameter glass tubing holding a paper-pulp filter.

Full details of the treatment for each spot test are given in the original paper, while the basic reagents used are as follows: nickel — dimethylglyoxime; chromium—diphenyl carbazide; manganese—sodium bismuthate; molybdenum—potassium ethyl xanthate; tungsten—ammonium persulphate - oxalic acid; aluminium—aurin tricarboxylic acid; copper—α-benzoin oxime; lead—dithizone; titanium—chromotropic acid; cobalt — α-nitroso-β-naphthol; selenium — bromine; vanadium—diphenyl carbazone.

A warning is added concerning the difficulty attached to interpreting these spot tests quantitatively. This arises primarily because, from one type of steel to another, the chemical attack of the stripping agent varies, so that while it *may* be possible to compare two steels of precisely the same type, if the sizes of the reagent drops are carefully controlled, comparison of two steels of different types is undoubtedly very hazardous.

1. B. S. Evans and D. G. Higgs. *Analyst*, 1945, **70**, 75.

New Strength Aluminium Alloys

DURING recent years a number of high-strength aluminium alloys, both cast and wrought, have been developed. The engineering properties of one of these, OH 38, a strong cast alloy requiring no heat-treatments, has been presented recently by A. J. Matter.[1] This alloy, which contains aluminium, zinc and manganese, can be produced from secondary aluminium alloy. Its mechanical and other properties are given in Table 1. As regards working properties, it requires no hardening or annealing, and consequently no straightening operations. It machines easily and threads can be cut to sharp edges that will not rupture or split. Corrosion tests have shown it to be immune to the corrosive effects of carbon tetrachloride, lactic acid, ammonia and vinegar. Castings of the alloy polish to mirror-like brilliancy and can be anodized or plated with chromium, nickel or tin. Service uses to which this alloy has been put are for hand-locking nuts, rocker arms, radio parts and automobile parts.

TABLE 1.—ENGINEERING PROPERTIES OF ALLOY

Tensile strength	17 to 19 tons per sq. in.
Elongation	2–4%
Brinell hardness	98·5 (average)
Specific gravity	2·88
Coeff. of linear exp. per deg. F. between 70° and 300° F.	$13·5 \times 10^{-6}$
Conductivity	33⅓% of copper's
Good bearing and wearing qualities	
Non-magnetic.	Non-sparking.

As a result of a number of years of intensive production, a new alloy, 75 S, has been put into commercial production, and its properties are discussed by J. A. Nock.[2] This alloy contains approximately 90% aluminium with magnesium, zinc and copper as its major constituents, and is commercially available in the form of alclad sheet and extruded shapes. In Table 2 are given the mechanical properties of this alloy, together with the values for Young's modulus of elasticity, the shear strength, hardness and endurance limit.

The annealing of 75 S products is accomplished by a two-stage treatment which produces maximum workability and freedom from age-hardening. The first, a recrystallization stage, consists of heating for 2 hours at 415°–455° C. and cooling in air. The second, a precipitation stage consists of soaking the material for 4 hours at 230° C. Alclad 75 S can be heat-treated over the temperature range 460°–500° C. followed by rapid quenching in cold water, but a lower maximum temperature of 470° C. is recommended for extrusions. Products of 75 S alloy age-harden at room temperature subsequent to solution heat-treatment, and continue to harden for an indefinite period. Ageing, however, can be retarded by storing at sub-normal temperatures. An elevated temperature precipitation treatment (artificial ageing) is required to produce the fully heat-treated temper. The recommended treatment is 20 to 24 hours at 120° C., and maximum strengths will be attained if the artificial ageing is started within two hours after quenching or else delayed for two days.

The strengths of freshly quenched Alclad 75 S sheet suggests the use of material in this condition for many forming operations, as the strengths show no increase for at least 1 hour. In difficult forming operations requiring several softening treatments, it is recommended that a solution heat-treatment be used for the final softening operation in order to avoid the distortion usually encountered in the heat-treatment of formed parts and to eliminate difficult straightening operations.

Results from a large number of accelerated corrosion tests on alloy 75 S and from limited exposure to natural environments lead to the conclusion that Alclad 75 S—T sheet has a resistance to corrosion similar to that of Alclad 24 S—T, but is less affected by slower quenching rates and diffusion, particularly in the thin gauges. Extrusions of 75 S—T have a resistance to general corrosion and stress corrosion-cracking comparable to that of 24 S—T extrusions. Alclad 75 S—T sheet can be satisfactorily spot welded with the equipment ordinarily used in the aircraft industry.

Another new clad material, R 301, consisting of a high-strength core alloy, clad with a medium-strength corrosion-resistant alloy, is dealt with by T. L. Fritzlen and L. F. Mondolfo.[3] The core

TABLE 2.—MECHANICAL PROPERTIES OF 75 S ALLOY

Property	Sheet				Extrusions	
			Alclad 75S—T			
	Alclad 75 S—O	Alclad 75 S—W	0·016–0·039 in. thick	0·040–0·249 in. thick	75 S—O	75 S—T
Tensile strength—Tons/sq. in. ..	16·0	23·0	38·0	38·5	20·0	44·0
Yield-point—Tons/sq. in.	7·5	10·0	33·0	33·5	10·0	40·0
Elongation (% on 2 in.)	16	20	10	12	12	10
Mod. of elast.—Lb./sq. in. ..	—	—	10,400,000	10,400,000	—	10,400,000
Shear-stress—Tons/sq. in.	—	—	23·0	25·0	—	23·5
Brinell hardness	—	—	—	—	—	145/160
Rockwell B. hardness	—	—	—	—	—	85/95
Endurance limit—Tons/sq. in. ..	—	—	—	—	—	11·25

TABLE 3.—TYPICAL PROPERTIES OF R 301

Property	R 301—O	R 301—W		R 301—T	
	All gauges	0·039 and lighter	0·040 and heavier	0·039 and lighter	0·040 and heavier
Tensile strength—Tons/sq. in. ..	12·5	29·5	30·5	33·0	34·5
Yield strength, 0·2%—Tons/sq. in..	5·0	19·5	20·5	30·0	30·5
Elongation (% on 2 in.)	22	19	20	9	9
Rockwell E hardness	—	—	—	105	100
Compressive yield-strength—Tons/sq. in.	—	—	—	31·0	31·0
Shear-strength—Tons/sq. in. ..	—	—	—	21·5	21·5
Endurance limit—Tons/sq. in. ..	—	—	—	6·25	6·25
Modulus of elasticity—Lb./sq. in. ..	10,300,000	10,300,000	10,300,000	10,300,000	10,300,000
Density,—Lbs./cu. in.	0·101	0·101	0·101	0·101	0·101

1 *Metals and Alloys*, 1944, vol. 20, No. 3, pp. 643–644.
2 *Metals and Alloys*, 1944, vol. 20, No. 4, pp. 922–925.
3 *Metals and Alloys*, 1944, vol. 20, No. 4, pp. 926–933.

alloy and the cladding alloy containing respectively 1·0 and 0·7% silicon, 4·5 and 0·4% copper, 0·8 and 0·5% manganese, 0·4 and 1·0% magnesium, 1·0 and 0·7% iron, 0·1 and 0·35% chromium, and 0·25 and 0·2% zinc. Cladding thickness varies with the gauge, ranging from 10% for a gauge of 0·024 in. or less to 2·5% for a gauge of 0·102 in. and heavier. The material is produced in three tempers annealed (O), solution treated (W), solution-treated and aged (T). Typical physical properties are given in Table 3.

In the heat-treatment of R 301, annealing to relieve hardening due to cold work is carried out at 350° C. and annealing to relieve hardening due to heat-treatment at 400° with slow cooling to 315° C. Solution heat-treatment is carried out 505° C., so as to reduce to a minimum diffusion of the copper into the cladding. When heat-treated from annealed material and naturally aged, the alloy shows mechanical properties somewhat lower than material heat-treated from cold rolled material, but, when material heat-treated from annealed material is aged to produce the T temper, that difference disappears completely and the mechanical properties will be as high as those of materials heat-treated from cold rolled. Natural ageing takes place at a rapid rate in the first 4 to 5 hours, reaching 90% of the final mechanical properties after 24 hours, after which ageing is slow, stability being attained in 10 days. For artificial ageing two treatments are recommended: (1) Ageing 18 hours at 160° C., when some forming has to be done after ageing; (2) ageing 6 hours at 175° C. when no more forming has to be done.

R 301 in the O temper forms better than most other high-strength alloys, and when freshly quenched, forms about as well as most other strong alloys in the annealed temper. In the W temper, it has also good formability and in many cases can be used to replace other high-strength alloys which must be formed in the O temper. In the T temper, R 301 forms approximately as well as other strong alloys in the same temper. The corrosion resistance of the alloy is high and compares favourably with that of other strong alloys. Very little difference exists between the corrosion resistance of the two tempers W and T, and neither temper is susceptible to intergranular corrosion. This new alloy can also be welded like other aluminium alloys by the established methods of oxyacetylene, arc, spot or seam welding, and can also be anodized and painted.

Designing Aluminium Forgings

By S. V. Malcuit

DURING the last few years there has been a hugh expansion in the use of aluminium forgings and the facilities for providing them. In the United States, from a total monthly production of a few hundred thousands of pounds of pre-war demand, requirements have risen in excess of 25,000,000 lb. per month, involving many thousands of entirely new parts, for each of which elaborate tooling has had to be provided. Aluminium alloy forgings, which offer the same general advantages over aluminium castings as steel or brass forgings offer over the same metals in cast form or as fabricated, riveted or welded structures, are produced for the most part on all types of standard forging equipment, steam hammer, board hammer, presses mechanical and hydraulic, etc.

In designing an aluminium forging, the first condition necessary is to determine the conditions which the part has to meet in service. In this respect loading, whether cyclic or static, is important, as is also the matter of stiffness or resiliency. Other important factors which have to be considered are the temperature conditions the part has to meet, the amount of wear or abrasion to which it will be subjected, and the corrosive conditions it will be exposed to. In some forgings, also, the ultimate use of the part may depend on such properties as electrical conductivity, thermal conductivity or thermal expansion, and these properties have to be taken into account.

In Table I are given the typical mechanical properties of nine aluminium alloys used for forgings. The strongest alloy in general use is 14 S, and is generally used where high strength and good resistance to corrosion are required. It is, however, difficult to forge, and when doing so, tolerances must be closely adhered to, and dies must be correctly proportioned to permit ready flow of the metal. Alloys 25 S and 17 S are used whereever moderate strength is required. Alloy 25 S is easier to forge than 17 S, but the latter has superior machining qualities and better resistance to corrosion. For the production of large and intricate parts of high mechanical strength, alloy A 51 S is generally used. Unusually severe corrosive conditions necessitate the use of alloy 53 S, while alloys 32 S and 18 S are used for such parts as pistons where strength at elevated temperatures is essential. Alloy 32 S has the lowest coefficient of thermal expansion, and alloy 11 S is the most easily machinable.

In order to place the flow lines of the metal to the best advantage and to suit particular requirements, forging stock must be positioned in the first forging operation. The exact positioning and the subsequent forging operations depend on the complexity and shape of the forged part and upon the direction of the important stresses on the part while in service. Once the design has been modified so that stresses are advantageously distributed, the detail of the design must be worked out. The designer must choose the forging plane which lies at right angle to the direction in which the reciprocating die travels. The parting line along which the lower and upper dies separate may be straight

TABLE I.—Mechanical Properties of Aluminium Alloy Forgings

Alloy	Minimum specification values					Typical values	
	Tension			Hardness Brinell	Shear Shearing strength	Fatigue Endurance limit	Density
	Yield strength	Ultimate strength	Elongation				
	Lb./sq. in.	Lb./sq. in.	% on 2 in.	500 kg and 10 mm ball	Lb./sq. in.	Lb./sq. in.	Lb./cu. in.
11 S—T	34,000	55,000	12·0	90	31,000	13,000	0·101
14 S—T	55,000	65,000	10·0	130	45,000	16,000	0·101
17 S—T	30,000	55,000	16·0	100	36,000	15,000	0·101
18 S—T	40,000	55,000	10·0	100	38,000	14,500	0·103
25 S—T	30,000	55,000	16·0	100	35,000	15,000	0·101
32 S—T	40,000	52,000	5·0	115	38,000	14,000	0·097
A 51 S—T	34,000	44,000	14·0	90	32,000	10,500	0·097
53 S—T	30,000	36,000	16·0	75	24,000	11,000	0·097
70 S—T	40,000	50,000	16·0	85	37,000	19,000	0·107

Metals and Alloys, 1944, vol. 20, No. 5, pp. 1319–1325.

and lie along the forging plane, or it may be made irregular to accommodate such parts as heads, lugs and webs. To prevent the forging from sticking to the die, a draft angle of approximately 7° must be provided. The size of radii and fillets should be as large as possible to assist the flow of metal through the change in direction. Where holes are to be punched in webs, they should start at the point of tangency of the fillet and the web. Sufficient metal should be allowed to compensate for shrinkage, mismatch and machining.

In order to distribute the stress in the design that has been proportioned according to experience and mathematical calculations, a stress analysis by strain measurements should be carried out. This is done by coating a full sized cast part with a brittle lacquer and then loading it so as to simulate the conditions of loading found in service, when cracks in the lacquer will indicate where stresses are of high order. Strain gauge readings should then be made of such regions, to determine the relationship between the load and the strain. Where stress analysis warrants changes, a part should be redesigned until the stresses approach a safe level with a minimum of stress concentration.

The Fused Salt Electrolysis for the Production of Metal Powders

By W. J. Kroll

THE recent wide-spread interest in powder metallurgy has suggested the production of relatively pure metal powders or metal-alloy powders by the fused electrolyte method. The three major metal products of the fused electrolyte industry are sodium, aluminium and magnesium. In each case the temperature of the fused salt bath is decidedly above the melting point of the metal produced. Accordingly all three metals are removed from the cells in liquid form. But when the melting point of the metal is above the bath temperature the metal will be deposited as small crystals or as dendrites. About 20 metals have been deposited in this form.

If fused salt electrolyses are carried out at a temperature above the melting point of the electrolyte but below that of the metal to be deposited at the cathode, a deposit of metal dendrites is obtained, at high cathode current densities, interspersed with electrolyte. This method has been used in the past with baths constituted of halides, borates, phosphates, etc., for the production of thorium, uranium, tantalum, cobalt, nickel, iron, copper, zirconium, silver, platinum, beryllium, molybdenum, tungsten, chromium, and aluminium. The advantages of this type of electrolysis are lower operating temperature and the resulting reduced power consumption; the winning of metals far below their melting point; and operating at relatively high current densities since the fine crystals covering the face of the cathode present a relatively large actual surface. Refining can be accomplished by using soluble anodes. The disadvantages are the difficulties encountered in removing the deposits from the cell, the elimination of the entrapped electrolyte, and the reaction of air and moisture with the bath.

Actual cell operation was studied. Smooth cathode deposits (sometimes in layers) have been obtained by earlier investigators at very low densities (below 4 amp./dm.^{2}). At higher densities, the crystals are oriented and grow into dendrites. The metals with a higher electrodeposition voltage tend to accumulate as salts in the bath. Upon increasing the current density, the metal constituents of these salts may also deposit at the cathode. This is due to the fact that the rate of deposition by the current may be faster than the rate of dissolution from the cathode by chemical equilibrium reactions. Upon interrupting the current, the chemical equilibrium-current, the chemical equilibrium reaction may proceed. This takes time, since alloys have formed on the cathode and solid metal diffusion retards the attainment of equilibrium. Intermetallic compound formation and depletion of certain ions next the cathode will also favour the codeposition of metals having a higher decomposition voltage. Eventually even the metals of the solvent or carrier salt, such as K and Na, will be codeposited.

The refining action at the anode is unpredictable since it depends on the decomposition voltage; as well as on the volatility of the compounds formed at the anode (such as $SiCl_4$); on the dissolution or stability of such compounds; on the rate of mixing of the bath; on the formation of intermetallic compounds at the cathode; and on the relative masses of metals and salts involved in the chemical equilibrium between bath and anode or bath and cathode. At low current densities the refining process may be better, due to the predominance of the chemical equilibrium over the electrolytic deposition equilibrium. The codeposited metal has time to react back with the bath at low current densities.

In the typical case of refining iron by anodic dissolution in an alkali chloride bath, carbon, sulphur, silicon and phosphorus are eliminated. The iron product obtained is very pure; soft, and free of nitrogen and hydrogen. It is dense, sharp-edged, free-running, and of grain size suitable particularly for powder metallurgy. In the electrolysis of a trivalent iron ($FeCl_3$) bath with insoluble carbon anodes, it is shown that the current efficiency is low due to back reactions of iron trichloride with the deposited iron sponge, forming iron dichloride, and due to the reformation of iron trichloride at the anode by part of the free chlorine. The electrolysis with an alloy anode is described, using for illustration a ferromanganese anode in a manganese chloride-containing alkali chloride bath. Iron and manganese are deposited and it is shown that chemical equilibrium conditions are not reached since the deposit has no time to react back with the electrolyte.

From *Electrochem. Soc.* Preprint, 87-5, 1945.

Transfer Car

A 110-ton side dump diesel-electric ore transfer car to be used in blast furnace and steel plant service has been developed by the Atlas Car and Mfg. Co. Cleveland Ohio U.S.A. It is powered by two diesel engines each rated 200 h.p. at 1,600 r.p.m. and each driving a suitable generator. Complete power equipment at each end of the car is synchronised electrically so that it can be fully controlled when operating on both power plants or on either one of them from the control station in either cab.

The Iron Age, January 25, 1945.

Quantitative Spectroscopic Analysis of Copper Alloys

By R. A. Wolfe and E. J. Jeml

AT the present time many industrial laboratories are using the spectrograph as a tool for making accurate and rapid analysis of metals for control purposes. In the main such analysis have been restricted to elements present only as traces or to alloying constituents with a concentration under 5%. A method has been investigated, the details of which are given, for the routine spectographic analysis of copper alloys in which a large number of the elements determined are at the 10% level, while an occasional analysis is made with contents as high as 25%.

During the initial stages of the investigation, several standard methods of spectroscopic procedures were tried and since it was necessary to analyse the alloys for elements at high concentration, it quickly became evident that only some type of a rigorously controlled spark source would give the necessary analytical precision. Two types of spark sources were used for the investigation. One was a standard high voltage source with synchronous rotating spark gap in series with the analytical spark gap, and the other a high voltage spark with an auxiliary air gap.

With copper alloys, sampling difficulties were considerable, due to the the presence of lead. Many methods of sampling were tried and the best technique was found to be that employing the use of large metal moulds designed to give pins ¼ in. by 1¼ in. and obtained when very little hot top was left in the mould. Pairs of electrodes were made by discarding ¼ in. of the tip of the pins opposite the hot top and sparking the two surfaces thus formed. Careful consideration was given to the shape and size of the electrodes used and the ends of the electrodes were prepared by grinding on a wheel of suitable composition and hardness. Flat surfaces were found to be more reproducible, the sharp edges being slightly rounded off. Excitation conditions in the spark were varied until a complete and uniform condition existed after pre-sparking. To find when the electrodes had been run in the spark sufficiently long to reach a stable condition, time of wait curves were determined under various conditions.

TABLE I.
REPEAT ANALYSIS FOR TIN ON STANDARD SAMPLE.

Micro-Photometric Reading, MM.		Log. $\frac{I_{SN3330}}{I_{CU3338}}$	Deviation of $\frac{I_{SN}}{I_{CU}}$ Log. From Mean
3338 Å, Copper	3330 Å, Tin		
297	335	0·35	0·003
340	370	0·34	0·013
351	380	0·35	0·003
356	382	0·34	0·013
360	380	0·37	0·017
345	376	0·36	0·007
317	355	0·38	0·027
327	356	0·30	0·053
358	389	0·39	0·037
336	371	0·39	0·037
311	348	0·36	0·007
326	361	0·36	0·007

The investigation was limited to copper alloys with two variable constituents. Analytical curves were prepared for tin in an 88–10–2 alloy with zinc constant at 2·75%, and for zinc in an 88–10–2 alloy with tin constant at 10%. In order to avoid having to make an extremely large number of analytical curves, based on a large number of possible alloy variations, a pair of correction curves for tin and zinc were worked out. With those curves the procedure for analysing unknown copper alloys containing tin and zinc consisted in taking spectrograms within the working range and determining the logarithms of the ratios of intensity of the selected tin, zinc and copper lines in the usual way. The apparent tin and zinc contents are then determined from their respective analytical curves and and true copper and zinc contents from the correction curves.

To test the method under investigation, a sample of 88–10–2 alloy was taken and analysed a great many times under carefully controlled conditions. The electrodes were ¼ in. rods about 1½ in. long. Flat surfaces were ground on each electrode between analysis and the edges were rounded off. A sparking time of 90 seconds was used with the spark circuit described using the auxiliary air jet. No air was blown through the analytical gap. Exposure time was 35 seconds on a contrast plate, and the plates used were calibrated from a nickel spark spectrum. Results obtained are given in Table I.

TABLE II.
COMPARISON OF VALUES OBTAINED BY CHEMICAL AND SPECTROGRAPHIC METHODS.

Sample	Tin		Zinc	
	Chemical	Spectrographic	Chemical	Spectrographic
1	7·30	7·38	4·30	4·28
2	7·53	7·61	4·45	4·36
3	9·69	9·58	2·23	2·27
4	9·83	9·81	2·47	2·56
5	10·25	10·14	1·87	1·76
6	10·06	9·88	1/98	2·07
7	12·36	12·05	1·15	1·15
8	12·77	12·74	1·43	1·43

A comparison of values obtained from several copper-tin-zinc alloys taken by chemical analysis and the spectroscopic method are given in Table II. The samples for the chemical analysis were in most cases taken very close to the surfaces sparked for the spectrographic analysis.

Bulletin Amer. Soc. Testing Materials, 1944, No. 129, Aug., pp. 45–51.

Lead Coatings on Steel

A Composite

LEAD coatings on steel have been developed during recent years, to save more critical metals, such as zinc, cadmium, chromium and tin. The lead or lead alloy coating possesses a high degree of workability, and withstands severe deformations without injuring the coating, due largely to the soft and ductile lead. Lead covered sheets solder easily and rapidly, with less than the usual consumption of solder, and since lead is an effective lubricant, it assists in die forming, stamping and spinning operations. Lead also does not corrode or weather through galvanic action when in contact with steel, and lead coated sheets spot weld fairly readily and take and hold paint and synthetic enamels without a priming coat and without any preparatory treatment. In a recent paper, H. A. Knight[1] describes lead coatings, both hot-dipped and electrolytic, and discusses their engineering properties and applications.

Coatings for steel either consist of pure lead or of lead alloyed with tin, antimony, silver and zinc. Silver enhances the bonding properties of the coating, tin provides wetting, while antimony gives increased hardness and better resistance to weathering under hot humid conditions. Such coatings are usually applied either by hot-dipping or by electrodepositing. Electrodeposition of lead is most

1 *Metals and Alloys*, 1944, vol. 20, No. 5, pp. 1296–1310.

successfully accomplished when plated from a lead fluroborate solution. Lead fluroborate solutions are preferred to flurosilicate solutions, because steel can be plated directly, while the silicate solutions require a preliminary copper coating. Fluroborate solutions are also more stable and deposits are finer. Lead puchlorate solutions have also been used for electrodeposition of lead.

Lead coatings vary in thickness from 0·0005 to 0·001 in., depending on the conditions to which they are to be subjected. The exact thickness required to give an impermeable coating depends to a certain extent upon the smoothness and cleanness of the initial surface and the structure of the lead deposit. As regards pin-holding if coatings are of the proper thickness, 0·0005 in., pin-holes are not inimical to corrosion resistance, coatings being self-healing due to the formation of some kind of lead-iron oxide compound. In fabrication, parts which are spot welded, riveted or hand soldered, can be coated and welded at the same time by simply tacking here and there and then coating with lead alloy. Lead-coated steel sheet can be easily spot-welded to ordinary steel, but cannot be spot-welded to another lead-coated steel sheet.

In a report[2] of Sub-Committee II of the American Society of Testing Materials Committee, B-8 on performance tests of lead-coated steels, H. A. Pray discusses the results obtained from samples of steel, copper-plated steel and copper, all plated with 0·00008 to 0·002 in. of lead from commercial sulphamate and fluroborate baths, and exposed at four stations representing industrial, rural, maritime and tropical conditions respectively. Panels of electroplated lead-tin-coatings on steel, hot-dipped lead alloy coatings, electroplated and hot-dipped zinc, intentionally damaged lead-coatings on steel and bare steel, were also exposed.

It was found that electroplated lead coatings exposed to the atmosphere first either darken in appearance or acquire a film of a white corrosion product. The first evidence of attack on the steel base appears in the form of pin-hole rust spots which may occur in groups or colonies, or which may be uniformly distributed over the panel surface. On further exposure, the pin-hole rust spots either disappear or merge into measurable rust areas. After 0·07 yr., panels with 0·00008 in. lead coatings had suffered early and severe rusting. No creeping of rust under the lead plating at the bare steel-lead interfaces on the damaged panels was found to occur at any station.

In an appendix[3] to the report of Sub-Committee VIII of the American Society for Testing Material Committee A–5 on field tests of metallic coatings, proposed specifications for lead-alloy (hot-dip) coated steel sheets and a tentative specification for lead-coating (hot-dip) on iron or steel hardware are given. The former specifications cover steel sheets, lead-alloy (2·5% tin) coated on acid or basic open-hearth steel, no weight of coating specified being given. In the Report itself, the progress of rusting and perforation of galvanized sheets and various plated hardware, structural shape, and tubular goods exposed since about 1926 at five locations in the United States are dealt with. Pin-holing and rusting in the lead-coated parts was found to vary at the various test stations.

2 *Amer. Soc. Testing Mats.*, 1944, Preprint No. 19 p. 8.

3 *Amer. Soc. Testing Mats.*, 1944, Preprint No. 9, p. 31.

The Sintering of Metal Powder—Copper

By C. J. Bier and J. F. O'Keeff

A STUDY of the mechanism of the sintering process in copper powder has been made covering the three variables affecting the sintering process, namely, briquetting pressure, sintering temperature and time in hot zone. Although strength and dimentional changes were observed, the investigation was primarily a metallographic study of the sintering process in copper. The copper powder used was made from reduced copper oxide and was typical of a grade used for the mass production of parts formed of metal powders.

The work on the effect of briquetting pressure was done at a constant temperature for a definite time ; the effect of sintering temperature was observed at a definite briquetting pressure with a constant sintering time ; and the effect of sintering time was examined at a definite briquetting pressure and sintering temperature. Sintering was done in open boats in an electric furnace in an atmosphere of cracked city gas at an air-gas ratio commonly used in the industry. The effect of sintering temperature was conducted with the furnace at 1,550°, 1,600°, 1,650°, 1,675°, 1,700°, 1,775°, 1,800°, 1,850°, 1,900° and 1,930° F. ; a sintering time of ½ hour at the temperature and a briquetting pressure of 60,000 lb./sq. in. Sintering time experiments were conducted at time intervals of 2, 4, 6, 8, 10, 20, 30, 40, 50 and 60 min. in the hot zone at a constant temperature of 1,750° F. and a briquetting pressure of 60,000 lb./sq. in. The effect of briquetting pressures was examined at pressures of 20,000 ; 40,000 ; 60,000 ; 80,000 ; and 100,000 lb./sq. in. at a temperature of 1,750° F. and a time of ½ hour for all samples.

At very low sintering temperatures or at very short times at the usual sintering temperatures, although no changes can be observed in the microstructure, there is a noticeable increase in tensile strength, elongation, density and dimensional change. As the sintering temperature is raised, or somewhat longer sintering times at higher temperatures are used, there is a marked increase in strength and other properties, and these changes are reflected in the microstructure by a spread of the bonding areas between particles as a result of the reduction of the oxide coatings by the atmosphere used in sintering. At still higher temperatures or longer times, a merging of adjacent particles takes place, resulting in further increases in tensile strengths, elongating density and dimensional change. Increasing the sintering temperature beyond 1,750° F. results in a large increase in grain size with a decided falling-off in tensile strength and other properties. Physical properties are also affected by the rate of heating of the sample.

In following the progress of the sintering process microscopically, a very narrow range of temperatures and times is reached within which an entirely new aspect of the microstructure appears. Up to this range, while reduction of the oxide coatings and merging of adjacent particles takes place, the structure is found to consist of a non-uniform mass of large and small particles. Within the narrow range of temperatures and times, 30 min. at 1,650° F. to 12 min. at 1,750° F., the structure became more uniform in regard to grain size, the identity of the large and small particles is lost, a general increase in grain size is observed, and the structure resembled that of a cast, wrought and annealed copper.

(*Metals Technology*, 1944, vol. 11, No. 7, and A.I.M.M.E. Technical Publication No. 1765, pp. 16).

The effect of briquetting pressure is most obvious in regard to porosity, increasing briquetting pressures decreases the porosity. The tensile strength, elongation and density are increased by increasing briquetting pressure, while the shrinkage decreases. The decrease in porosity probably accounts for the greatest part of the increase in the tensile strength. Microscope examination does not show significant grain-size differences among the sintered structures briquetted at the various pressures covered in the investigation.

Pickling Eliminated by Continuous Salt Bath Anneal

FULL annealing of brass cartridge cases on a continuous production heat-treatment basis is being accomplished with the longest continuous conveyer-type salt bath furnaces in operation in the United States. The cases come out of the heat treatment clean and bright and without the slightest trace of oxidation, so that pickling or other cleaning is not required following the anneal. Two Upton electric salt bath furnaces are employed, each capable of annealing 6,000 lb. (2¾ metric tons) of brass every hour. One gives the cases a full anneal at 980° F. (527° C.) before the nosing operation; the second furnace, after the nosing operation. Each of the two furnaces replaces 10 individual non-continuous furnaces previously used. They have paid for themselves in the first three weeks of operation.

Cases from the previous operation are lowered into the molten salt at one end of furnace pot and moved along through the molten salt at a rate of 60 ft. (18·30 m.) per min. A turn-around area at the opposite end permits the conveyer to bring the cases back through the pot for the additional time required for the full anneal. Three shell cases are suspended from each hanger of the conveyer. The annealing operation is entirely automatic, with the exception of the occasional shovel-full of salt required to replenish the salt lost through normal operation. The temperature of the bath is held within ±5° F. (2·8° C.) by automatic electric temperature control and signal alarm, should the range be exceeded.

The first furnace to be installed was equipped with removable covers with the idea that sufficient heat would be radiated from the salt to make them advisable. Salt is added through a small door in the covers at the turn-around end of the furnace. This furnace has been in continuous operation for 18 months and has been shut down only twice during that time for cleaning. The previous equipment required shut-down for complete cleaning every week. The second furnace to be installed is not equipped with covers since it has been found that the amount of heat lost is small. Uniform temperature throughout the entire bath is made possible largely through the placement of electrodes which enter the sides of the furnace instead of being thrust in from above the surface of the salt. These electrodes are so positioned that the transfer of current through the salt generates the heat at the bottom of the pot. The result is that a normal flow of heated salt circulates upwards, heating the entire contents of the bath evenly and uniformly. Electrodes of this type, which are not exposed to the combined action of salt and oxygen, appear to be capable of lasting almost indefinitely, since in the 18 months of continuous operation, no replacements have been required. This electrode design in conjunction with the full ceramic pot permits the use of ordinary commercial nitrate salts, further reducing heat-treating costs. Only 20% of the volume of salt used previously by the individual furnaces is now required.

The Iron Age, January 18, 1945.

Nickel and Inconel Thermo-Couple Protection Tubes

By E. W. Cline and A. M. Hall

THERMOCOUPLE protection tubes are made of various materials—refractory or metallic but usually the former—depending on the conditions of temperature mechanical service and environment of the thermocouple service. Metal thermocouple tubes are used primarily to protect the enclosed thermocouple against mechanical shock and secondarily to protect it from chemical attack by the surrounding medium gaseous or liquid. Nickel and Inconel protection tubes have been found effective for service in molten lead or salt baths and in reducing atmospheres; and in oxidising atmospheres where they are not required for corrosion reasons, they are useful if mechanical shock is involved. A metal protection tube is generally made of cold-drawn seamless tubing closed at the hot end by a welded-on disc of the same metal and is usually installed in the furnace in the vertical position to prevent sagging.

In Table I are given data on the service conditions and service life of nickel and Inconel protection tubes in a lead bath, a salt bath, and various reducing atmospheres. To obtain the best service from an Inconel protection tube in a lead bath, it should be thoroughly soft annealed prior to use to remove the danger of splitting from stress corrosion. Nickel protection tubes are found suitable for immersion in fused chlorides in the absence of sulphur for example in barium chloride at 1010 to 1100° C., or in sodium and potassium chlorides operated up to 980° C. Inconel is recommended where

Metals and Alloys, 1945. Vol. 21, No. 2, pp. 401–403.

TABLE I.—THERMOCOUPLE PROTECTION TUBE SERVICE.

Metal	Temperature Range	Composition of Gas %							Other Media.	Life.
		O_2	CO	CO_2	N_2	H_2	H_2O	S		
Inconel ..	370°–980° C.	—	—	—	—	—	—	—	Lead Bath	3–5 Years.
Inconel ..	790°–870° C.	—	—	—	—	—	—	—	Cyanide Carburising Bath	3–5 Years Intermittent Operation.
Inconel ..	980°–1260° C.	—	5·0	8·0	64·0	5·0	18·0	0·0008*	—	1–2 Years.
Inconel ..	980°–1200° C.	—	2·0	11·0	64·0	2·0	21·0	0·0008*	—	1–2 Years.
Nickel ..	540°–790° C.	2·0	—	11·0	66·0	—	21·0	0·0008*	—	5–10 Years.
Nickel ..	815°–1065° C.	—	5·0	8·0	64·0	5·0	18·0	0·0008*	—	1–2 Years.
Nickel ..	670°–790° C.	—	—	—	90·0	10·0	Nil	Nil	—	5–10 Years.

* 0·5 grains per 100 cu. ft. of gas.

there is danger that the salt bath may be contaminated by sulphur. Inconel is also satisfactory for use in cyanide baths used for the liquid case-hardening of steel.

Reducing atmospheres in which metal protection tubes are used are generated in open furnaces by the incomplete combustion of mixtures of low sulphur, natural gas and air. For such atmospheres nickel or Inconel is used, but as a rule nickel is preferred because the gas molecules have more difficulty permeating it to attack the enclosed thermocouple. Permeability is very important in considering reducing gas atmospheres because the carbon monoxide and hydrogen which are usually their chief components are extremely harmful to thermocouples. When precious metal thermocouples are used an inner protection tube of a refractory material is also used to increase the protection from gaseous attack. Nickel protection tubes are also used in the artificial bright annealing atmosphere which is a mixture of 90% nitrogen, 10% hydrogen. At higher hydrogen concentrations metal protection tubes are useless because of the great permeating power of hydrogen, but at 10% hydrogen by volume long life is obtained.

Where metal protection tubes are used, as in oxidising atmospheres to resist mechanical breakage, nickel tubes are preferred because the adherent, continuous and relatively impenetrable oxide coating which forms on the metal provides it with a long life and at moderate temperatures the life of the metal is almost indefinite. Inconel is used as a substitute for expensive high-temperature refractory protection tubes in dry heat applications up to 1260° C., where mechanical shock is likely. The refractory materials are brittle, especially when hot, whereas Inconel retains considerable stiffness and is ductile, and is also stiffer than nickel at all temperatures.

Characteristics and Properties of some Cast Chromium-Molybdenum Steels

By N. A. Ziegler and W. L. Meinhart

CHROMIUM - MOLYBDENUM steels containing 5 to 9% chromium and 0·5 to 1·0% molybdenum, commonly used in castings for corrosion and elevated temperature service, are noted for their thermal sluggishness, which makes them susceptible to cracking during welding and heat-treatment. As more detailed data on alloy steels of this type have been desired, thermal analysis, microexamination, determination of physical properties and welding experiments have been carried out on a series of steels containing 0·05 to 0·30% carbon, 2·5 to 9% chromium and up to 1·5% molybdenum.

The steels were prepared in a 200-lb. basic lined high-frequency induction furnace and cast into blocks. Each block, heat-treated by normalising at 955° C., air cooled from 840° C., and tempering at 680° C., was tested for tensile properties and Charpy impact resistance, the latter test being carried out at room temperature and at −31° C. Most of the compositions were also subjected to thermal analyses, using a differential dilatometer and each dilatometer specimen was subjected to a micro-examination. Welding tests were made on all experimental castings after they were given the same heat-treatment as the test specimens. In all, 72 compositions were tested, divided into four main groups, containing 2·5, 5, 7 and 9% chromium. Each group, in turn, was sub-divided into five sub-groups, containing 0, 0·25, 0·5, 1·0 and 1·5% of molybdenum, and each sub-group was composed of three or more steels containing progressively higher carbon from theoretical zero and up to about 0·3% carbon.

Results obtained from the various tests showed that thermal sluggishness and hardenability increased in proportion to the percentage of carbon, chromium and/or molybdenum. The steels tested remained relatively soft after slow cooling (3·2° C. per minute) from the austenitic state regardless of the chromium and molybdenum contents, provided that carbon did not exceed 0·07%, and this means that there was no suppressed transformations developed. Fairly cooling (50° C. minute) did not cause any appreciable hardness increase in steels containing up to and including 2·5% chromium with and without molybdenum, and up to 5% chromium in the absence of molybdenum. Most of the 5 to 9% chromium low-carbon steels, however, were somewhat hardened by rapid cooling, as indicated by the appearance of a split transformation in the thermal cooling curves.

The 2·5 to 9% chromium steels in the 0·13 to 0·2% carbon range and with less than 0·5% molybdenum did not develop suppressed transformations and remained relatively soft after cooling from the austenitic state. Rapid cooling, however, made them fairly hard. Increasing the molybdenum content to the range of 0·5 to 1·5% resulted in slightly greater hardness on slow cooling and pronounced hardness and suppressed transformations on rapid cooling. When the carbon content exceeded 0·2%, these steels had marked suppressed transformations and became hard after rapid as well as after slow cooling from the sustenitic state.

Both chromium and molybdenum formed gamma loops in their equilibrium diagrams with iron and, as a result their combined effect on a given steel, above certain amounts was to make it at least partially ferritic at any temperature. With increasing amounts of carbon higher percentages of chromium and/or molybdenum steel became incompletely austenitic at the heat-treating temperature and in this state contained a certain amount of residual alpha ferrite which increased with further increases in the chromium and/or molybdenum contents. On cooling from the heat-treatment temperature, this residual ferrite acted as nuclei for the transformation of the austenite thus making such a steel less thermally sluggish.

When chromium and molybdenum contents were constant, increasing carbon was found to increase very considerably the strength and hardness and to reduce ductility and impact resistance, and these changes were particularly pronounced when carbon increased from about 0·07 to 0·15%. With chromium and carbon contents constant, an increase in molybdenum from 0 to 1·0% increased the strength and hardness and reduced the ductility and impact resistance. Further increase in molybdenum reduced the strength and hardness with a possible increase in ductility and impact resistance due to the appearance of the residual alpha ferrite. When carbon and molybdenum were constant, an increase in chromium from 0 to 6% increased the strength and hardness and reduced the ductility and impact resistance. Further increase in chromium up to 9% reduced strength

American Society for Metals, 1944. Preprint No. 34, pp. 1–39.

and hardness and somewhat increased ductility and impact resistance. The latter effect was particularly pronounced in higher molybdenum steels due to the presence of residual alpha ferrite caused by the combined influence of chromium and molybdenum.

The welding experiments showed the effect of thermal sluggishness and suppressed transformations on hardness distribution and internal strains in the weld metal and affected zones. They also indicated that a stress relieving treatment after welding was not nearly so essential in steels of the type tested, which contained under 0·15% carbon, as in those of a higher carbon content.

Fork Truck for Annealing Oven

By B. Gapinski

IT formerly required a full day for a complete cycle of loading and unloading an annealing oven at a foundry using a steam-powered oven charger. This cycle has been cut to just half that time by using a standard fork truck of 4,000 lb. (1,800 kg.) capacity. The annealing pots have to be stacked four deep six abreast across the oven from end to end. Formerly the work was done by a steam-powered, steel-wheeled, rear-end control oven charger. The operator must arrive two hours in advance of charging to fire-up and get up steam. Loss of steam and low truck speed proved inefficient. The rear-end control compelled the driver to remain in the end of the oven, picking up the last two rows of pots for 10 to 15 sec. in 600° F. (320° C.) heat on each trip. The fork truck that replaced the old equipment is gasoline-powered, with low overall height and controlled by the operator in the centre of the machine directly behind the load. The operator is protected from direct heat from the pots by a sheet steel shield mounted behind the forks. The low height permits operation against the oven side walls. The driver's position near the load gives quicker and more positive control of operations. The vehicle's speed means minimum time inside the oven. By use of a shovel attachment and an additional pair of forks, the truck can handle new and spent sand, core boxes, moulds and other items.

Metal and Alloys, January, 1945.

Alumina from Clay by Sulphuric Acid Process

By J. H. Walthall

ALUMINA has been made from clay on pilot-plant scale by a sulphuric acid process. From it has come metallic aluminium which compares favourably with that made from Bayer process alumina. Several of the operations that were developed should find other industrial applications, among which are the production of iron-free aluminium sulphate and of sulphuric acid from gases of relatively low sulphur dioxide content. The process consists of treating calcined clay with sulphuric acid to form aluminium sulphate; the iron and silica impurities are removed from the solution; the solution is evaporated to dryness. The aluminium sulphate is dehydrated and then decomposed to form oxides of sulphur, which are recovered as acid for re-use in the process, and the alumina from which impurities are removed by leaching.

Techniques were developed to overcome the previously recognised difficulties of acid processes for production of alumina. Iron was removed from the aluminium sulphate solution by treatment with manganese sulphate and ozone, and alternately by electrolysis. Indications are that the electrolytic and chemical methods are so nearly equal that a choice between them should await a thorough technical and economical appraisal. The corrosive solutions were concentrated by submerged combustion. Silica was removed from the partially concentrated solution by treatment with clay residue. The sulphur trioxide was hydrated and recovered as acid by electrostatic precipitation; the sulphur dioxide was absorbed in weak sulphuric acid and oxidised to sulphuric acid by aeration in the presence of a catalyst. In certain steps alternate methods were developed.

Data regarding these developments and other unit operations in the process demonstrate that the process operable and that the production of alumina suitable for reduction to aluminium is technically feasible. They also gave data and other information from which an economic evaluation of the process could be made. Recommendations regarding materials of construction for a commercial-size plant were given as well as a quantitative flow-sheet for a plant having a capacity of 162 tons of alumina per day.

From a paper before the American Institute of Chemical Engineers, St. Louis, November 20, 1944: *Chemical and Metallurgical Engineering*, January, 1945.

Chilled Iron with Tellurium

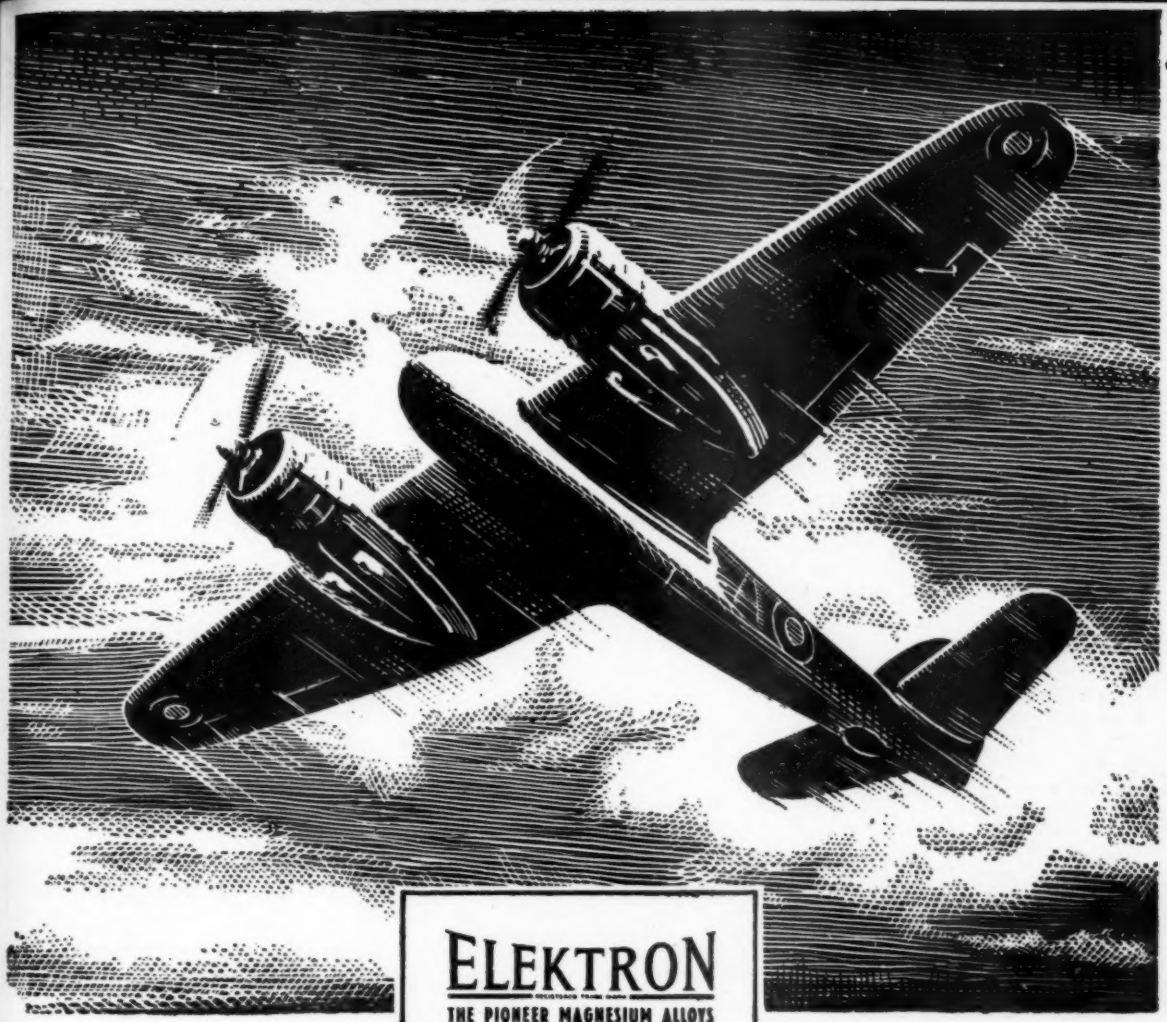
TELLURIUM is such a potent chill-producing alloying element that very minute additions of it to cast iron will increase the chill depth, hardness and resistance to certain types of chemical attack. Tellurium additions, such as solid metal or powder, usually range from 0·001 to 0·05%, but since its potency declines with time, it should be added just before pouring. It may also be added by painting or spraying the dry powder to the mould-metal interface, or by applying it to a pattern from which it is transferred to the mould surfaces. A localised chill can be obtained by coating specific areas of the mould surface with tellurium-bearing washes. With the tellurium content of a mould wash varying from 0·01 to 0·05 gm. per sq. in. (0·0016 to 0·0008 gm./cm.) the chill depth varied from $\frac{1}{16}$ to $\frac{3}{16}$ in. (1·6 to 4·8 mm.) on an iron containing 3·20 C, 1·70 Si, 0·50 Mn, 0·07 S, 0·20 P and 0·50% Cr.

In a test for resistance to acid attack, Long and Krause reported the following losses in grams after 26 hours' exposure for an untreated, unchilled gray iron and for a tellurium-treated gray iron :—

	2% H_2SO_4	2% HCl
Untreated, unchilled	4·16	5·51
Tellurium-treated	3·08	2·29

According to Smalley, an iron containing 3·1 C, 1·2 Si, 0·7 Mn, 0·095 S and 0·2% P, as cast against a metal chill with a resulting chill depth of $\frac{3}{8}$ in. (9·5 mm.) deep and a Brinell hardness of 415. An addition of 0·002 5% Te increased the chill depth to $1\frac{1}{2}$ in. (38 mm.) and the Brinell hardness to 460. The hardness increased to 469 with 0·005% Te, and 477 with 0·10%. Additions of copper, nickel and molybdenum with tellurium produce a hardness and toughness unobtainable with tellurium alone.

Condensed from J. S. Crout, *Foundry*, November 1944; *Metals and Alloys*, January, 1945.



ELEKTRON

THE PIONEER MAGNESIUM ALLOYS

The proved ELEKTRON alloys are produced in all forms by the most accomplished founders and wrought material manufacturers. ELEKTRON used in aircraft means increased carrying capacity.

Sole Producers and Proprietors of the Trade Mark "Elektron": MAGNESIUM ELEKTRON LIMITED, Abbey House, London, N.W.1 ● *Licensed Manufacturers, Castings and Forgings*: STERLING METALS LIMITED, Northby Road, Foleshill, Coventry ● *Castings*: THE BIRMINGHAM ALUMINIUM CASTING (1903) COMPANY LIMITED, Birmid Works, Smethwick, Birmingham ● J. STONE & COMPANY LIMITED, Deptford, London, S.E.14 ● *Sheet, Extrusions, Forgings & Tubes*: JAMES BOOTH & CO. (1915) LIMITED, Argyle Street Works, Nechells, Birmingham, 7 ● *Sheet, Extrusions, Etc.*: BIRMETALS LIMITED, Woodgate, Quinton, Birmingham ● *Suppliers of Magnesium and "Elektron" Metal for the British Empire*: F. A. HUGHES & CO. LIMITED, Abbey House, Baker Street, London, N.W.1

Impurities in Aluminium Alloys

IN order to determine the effect of small quantities of nickel and iron, separately and together, upon the properties of aluminium-copper-magnesium alloy sheet, a series of tests was made of various fabricating and heat treating procedures with controlled alloy compositions. The alloy had a nominal composition of 4% Cu and 0·60% Mg, balance aluminium. Iron-plus-nickel contents were tried to a maximum of 1%. Experimental ingots were hot-rolled longitudinally, annealed and cross-rolled, then cleaned chemically and annealed and cold-rolled. None of the alloys showed appreciable surface cracking, and edge cracking was slight. There was no noticeable tendency for iron or nickel to increase edge cracking. The effect of iron and nickel upon work-hardening was unappreciable.

Iron and nickel showed no effect upon annealing characteristics, but the effect on age-hardening was marked. The greater part of the age-hardening occurred between 1 hr. and 24 hr. after quenching, and in all the alloys the period of rapid age-hardening was followed by one of slow hardening, continuing to a slight degree over the full three-months period of test. Both the rate and extent of age-hardening were reduced progressively with increasing content of either nickel or iron, and when both are present, the effect is approximately additive to 1%. Tensile strength was progressively lowered by iron or nickel, with iron having the greater effect. A total iron-plus-nickel content of 0·75% caused the proof stress value to fall below specification requirements. The effect upon ductility was negligible.

Examination of the microstructure of the alloys was carried out to determine the extent to which intermetallic compounds remained unabsorbed after heat treatment. The shape and distribution has an important effect upon the bending qualities of the material. It was found that unabsorbed material was roughly proportional to the total iron-plus-nickel content. However, with increase of unabsorbed material there was a general refinement of particle size, so that relatively small deterioration of bending quality occurred. Corrosion resistance was tested in the conventional salt spray apparatus, and by an accelerated test using sodium chloride-hydrogen peroxide. Results showed that both iron and nickel in the higher proportions produced some slight increase in corrodibility, but that the change, in comparison with variations between alloys of different types, was of no practical importance.

Condensed from M. Cook & R. Chadwick, *Journal of the Institute of Metals*, September, 1944; *Metals and Alloys*, January, 1945.

Heat Treating Large Pieces

THE Southwest Welding and Manufacturing Co., in Alhambra, California, U.S.A., specialises in massive construction of alloys and non-ferrous metals that require a variety of stress reliefs, normalisers, and heat treatments. The stress relief furnace is 16 ft. wide, 16 ft. high and 50 ft. long (4·90, 4·90 and 15·25 m., respectively). It is externally fired with natural gas from two combustion chambers. If necessary, the fuel can be changed to oil within a few minutes. The combustion chambers and integrating hot gas duct are made of refractory brick; the balance is of 3½ in. (9 cm.) precast lightweight blocks. A 50,000 lb. (22 2/3 metric ton) charge can be heated from cold to 1,200° F. (650° C.) in this car-type furnace in 8 hrs. The 16 ft. × 16 ft. × 25 ft. (4·90, 4·90, 7·60 m. respectively) heat-treating furnace can also be fired with either gas or oil. The 10 in. (25 cm.) walls are of lightweight refractory brick, while the hearth and floor are of refractory brick. The furnace is direct-fired and can be heated to 2,200° F. (1,200° C.). The heat-treating and stress relief furnaces are separated by a water-cooled middle wall made of a monolithic refractory casting, 9 in. (23 cm.) thick in the main centre panel and 12 in. (30 cm.) thick on the sides. This wall can be removed to join the two furnaces if articles over 50 ft. (15·25 m.) long are to be stress relieved. Articles can be moved out of the furnace to the quench within 3 min. The water quench provides 1,500 g.p.m. (57 hl.p.m.) at 135 p.s.i. (9·5 kg./cm.²) and can remove 1,000,000 B.Th.U. (250,000 kg.cal.) per min. on 320 sq. ft. (30 m.²). A jet positioning device makes the quench very flexible. For alloys such as stainless, the quench can decrease the temperature from 2,200° F. (1,200° C.) to under 700° F. (370° C.) in less than 3 min.

Condensed from G. E. Stedman, "Industrial Gas," October, 1944, *Metals and Alloys*, January, 1945.

Chilling Aircraft Parts to make them Grow

A FREEZING process that causes under-sized engine parts to "grow" is now in regular use at the Ford Motor Co's plant at Dearborn, Michigan, U.S.A. The new treatment, which greatly reduces waste, is being applied to under-sized articulated rod pins for air-craft engines. Experiments with other parts have yielded promising results. Pins that are 0·0002 or 0·0003 in. (0·005 or 0·0075 mm.) under size are brought to the required dimensions by cooling them to a temperature of minus 313° F. (minus 192° C.), by packing them in liquid air. The actual "growth" takes place on the outer surface of the metal subjected to the cold-treatment because of a change in the physical properties of the austenite present in the steel. (When steel is subjected to extreme heat or extreme cold, the austenite changes into martensite, a constituent that has greater bulk.) Hence, the dimensions of the part treated are increased. Many thousands of pins have been salvaged through the cold-treatment.

Machinery, January, 1945.

Magnesium Airplane Wings in Training Planes

AIRPLANE wings fabricated entirely from magnesium have been used for more than a year on U.S. Navy training planes, according to a paper by J. C. Mathes of the Dow Chemical Co., read before the recent annual meeting of the American Society of Mechanical Engineers. While the use of magnesium alloys in aircraft has increased during the war to such an extent that the average fighting plane now utilises approximately half a ton of this metal, the adoption of all-magnesium wing structures represents a new departure in aircraft design. The speaker expressed the opinion that the success with these wings foreshadows all-magnesium airplanes in the near future. Since magnesium is considerably lighter than aluminium, several hundred pounds in plane weight can be saved by the utilisation of magnesium and air transport planes built from magnesium would be able to carry that many more pounds of pay load. It has been estimated that the saving of one lb. (450 gm.) in plane weight is worth $100 a year in a commercial plane.

Machinery, January, 1945.